BRITISH WELDING JOURNAL

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By C. G. Mackin and N. Y. Pulchak

Determination of Hydrogen in Weld Metal

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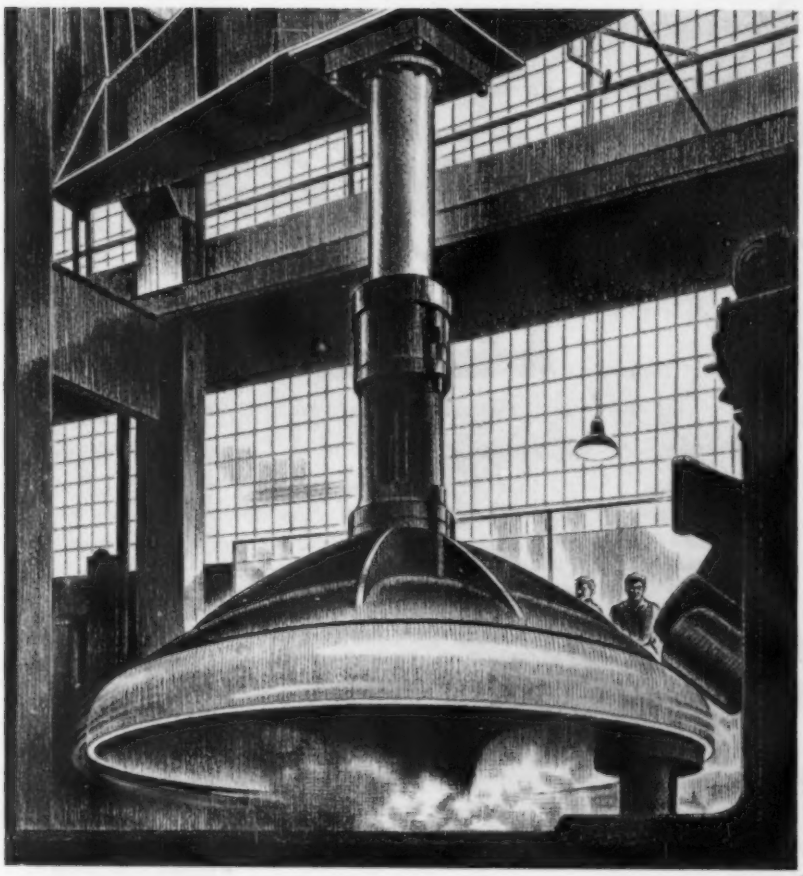

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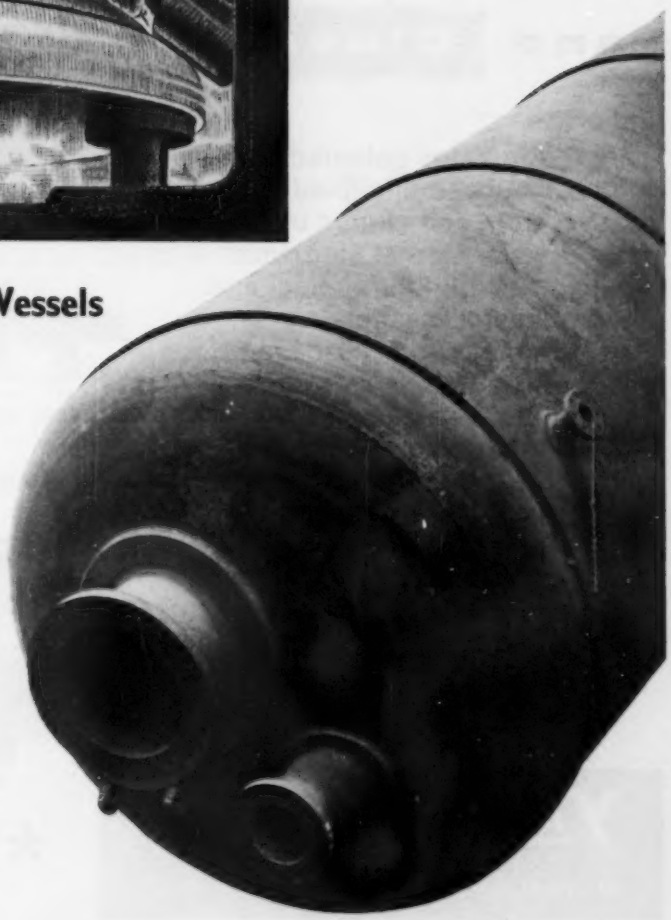

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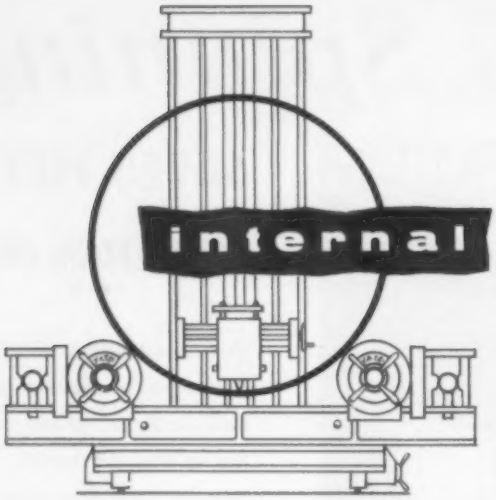

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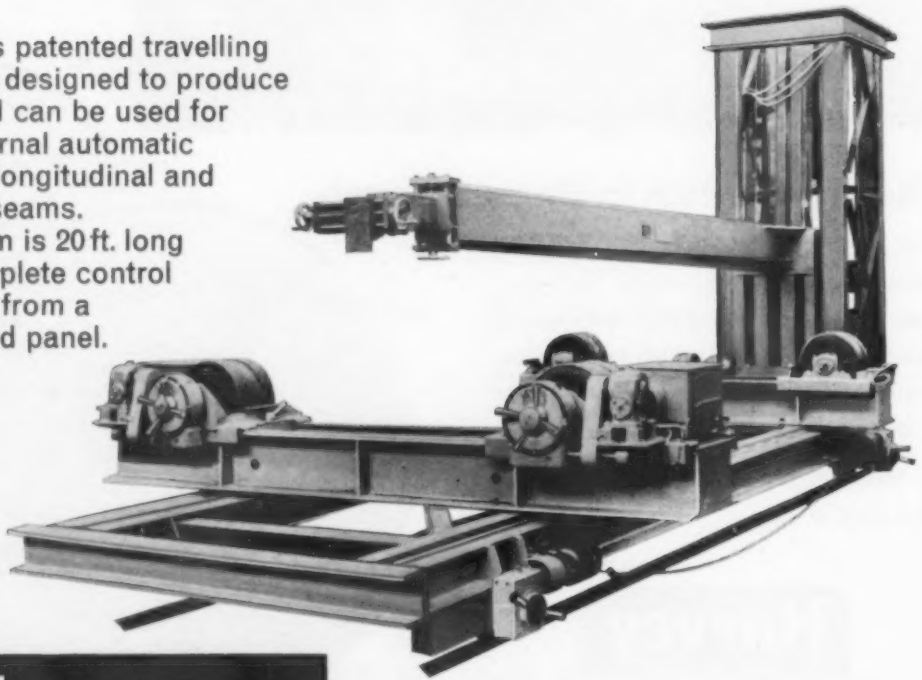

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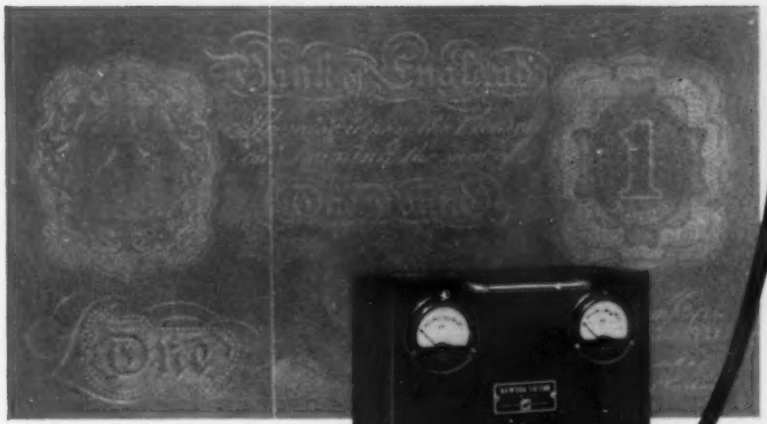

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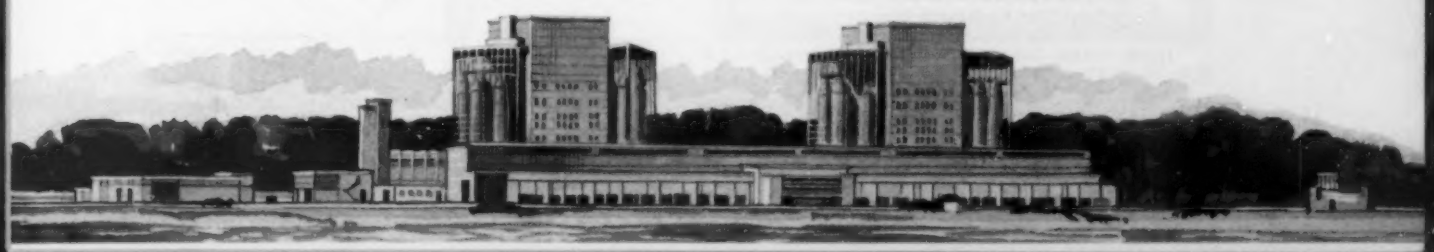
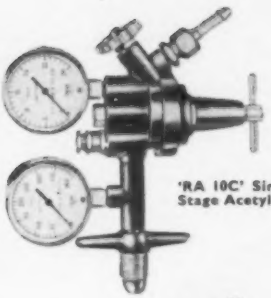


'TSS' Two Stage Oxygen with Hexagon Nut Cylinder Fitting



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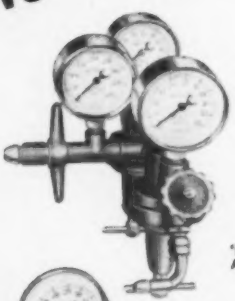

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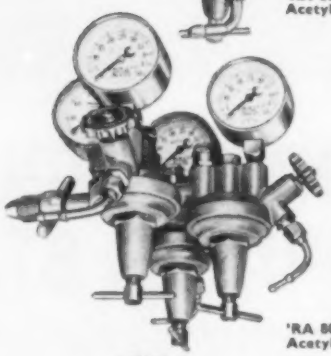

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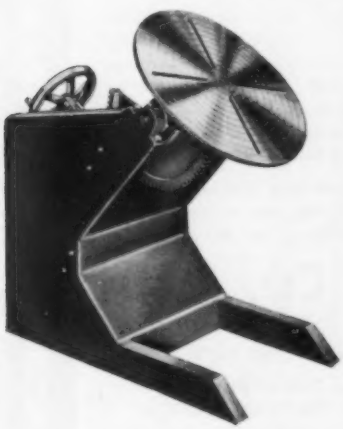

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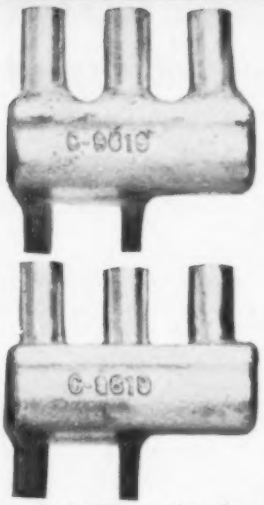
A high melting silicon bronze rod was used, with three fusion welds, to obtain suitable density, strength and corrosion resistance. Top welders did the work to overcome the problems of distortion, warping and keeping the pipes parallel.

The manufacturer was satisfied (Photo A) until Eutectic's Technical Representative recommended EutecRod 1801, a thin flowing high silver alloy, used together with Eutector Flux 1801.

The three tubes were pre-fluxed with Eutector Flux 1801 and set up simultaneously. A large torch tip with a broad flame was used to heat the work and when the flux liquified, EutecRod 1801 was fed into the joint. Continuous motion of the torch, and the superior capillarity of the alloy, prevented localised overheating and distortion, allowing uniform fillets to form around the entire circumference of each tube (Photo B).

Due to the extreme ease of application of EutecRod 1801, perfect joints were produced. Filler alloy was reduced by 56 grammes to the 6.2 grammes of EutecRod 1801 required by the three joints. The joints showed superior strength, corrosion resistance and "eye appeal". EutecRod 1801 reduced working temperature by at least 300°C, and rejects were diminished.



A.

B.

EutecRod 1801, a thin flowing high silver content alloy with medium plastic range, is one of the lowest melting but strongest of all silver bearing alloys, with an ultimate tensile strength of 40 tons/sq. in. Joints are ductile, and leakproof, offering exceptionally good corrosion resistance and electrical conducting properties.

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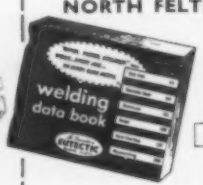

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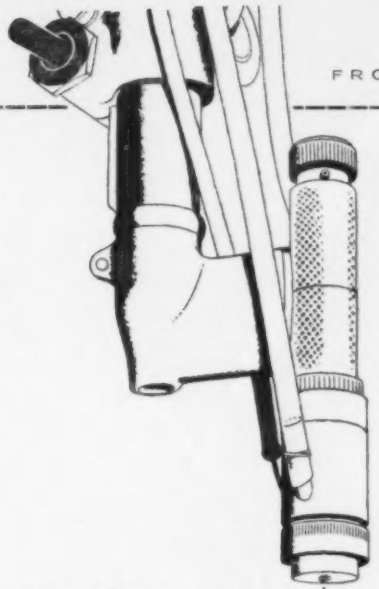

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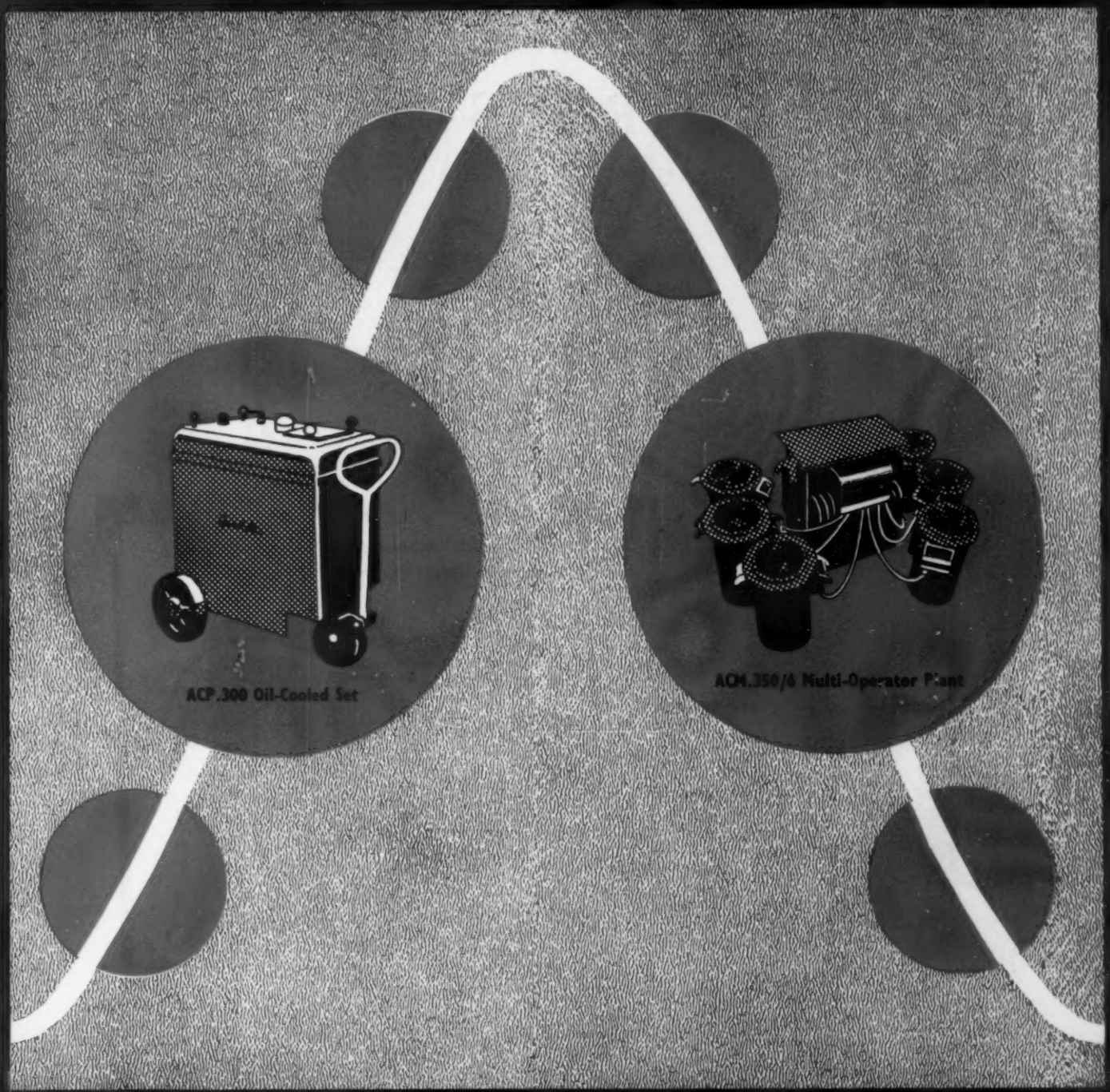

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BK 34	2400	355	—
BK 34A	2400	355	Integral
BK 34B	2400	355	Clamp on
BK 42	600	56	—
BK 42A	600	56	Integral
BK 42B	600	56	Clamp on
BK 66	300	22·4	—

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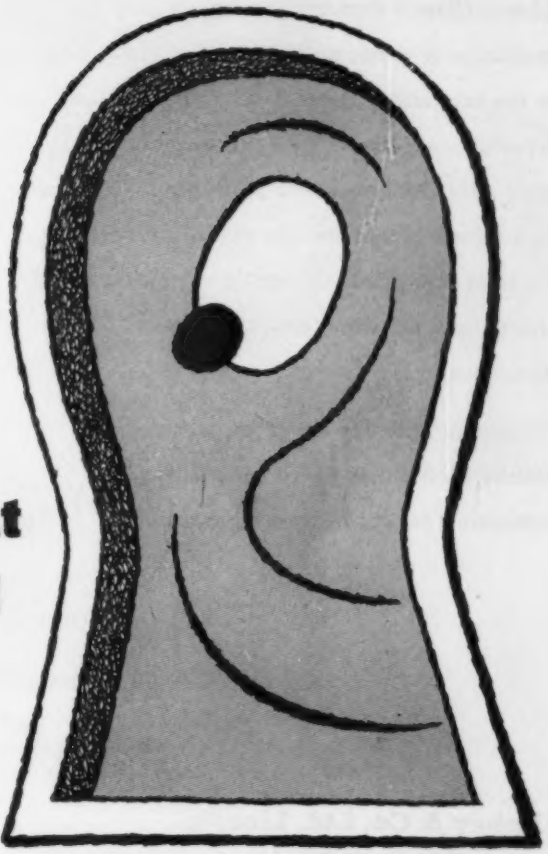

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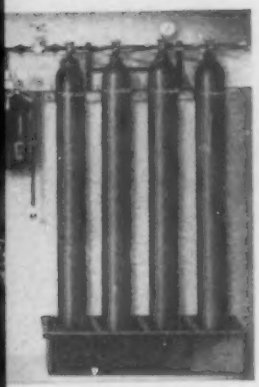
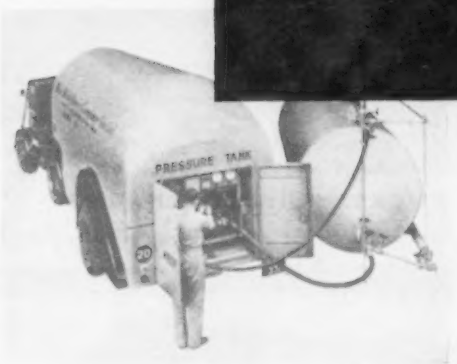
Photograph illustrating the FUSARC/CO_2 process by courtesy of Quasi-Arc Limited





The use of low pressure CO_2 gas as an inert shield in the welding of steels is increasing. CO_2 is not only cheap — it enables high quality welds to be made in the shortest time.

Where automatic welding machines have been specially modified to use this technique, the simplicity, ease of slag removal and high speed of welding have been found most impressive. On one differential housing, for example, welding time has been reduced from 105 to 23 seconds.

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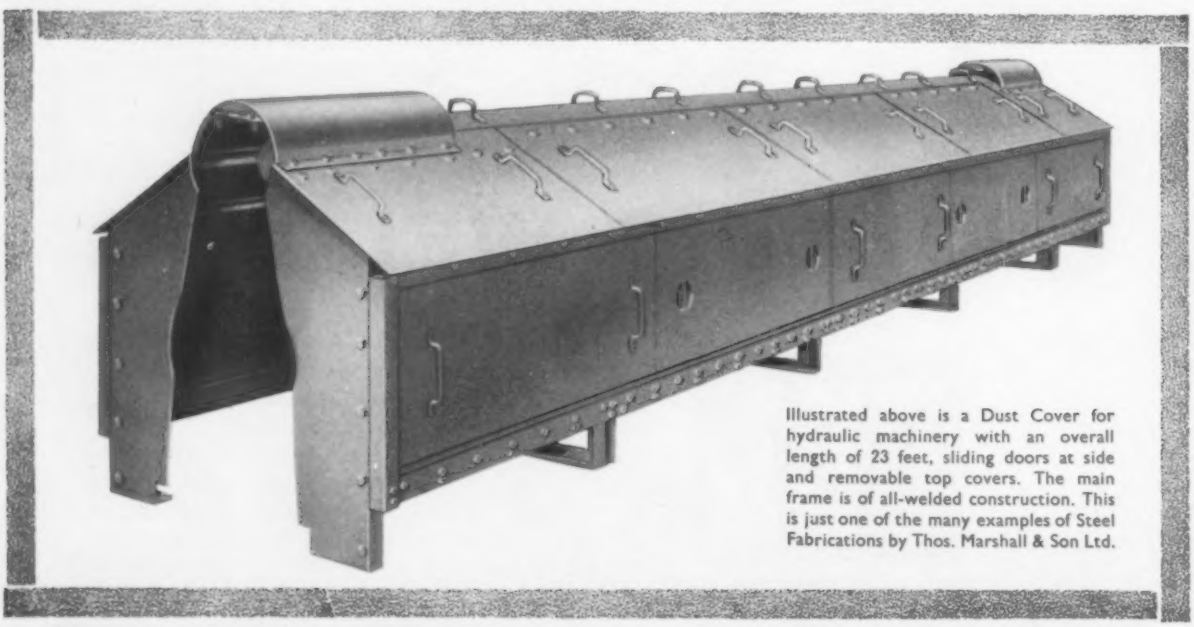

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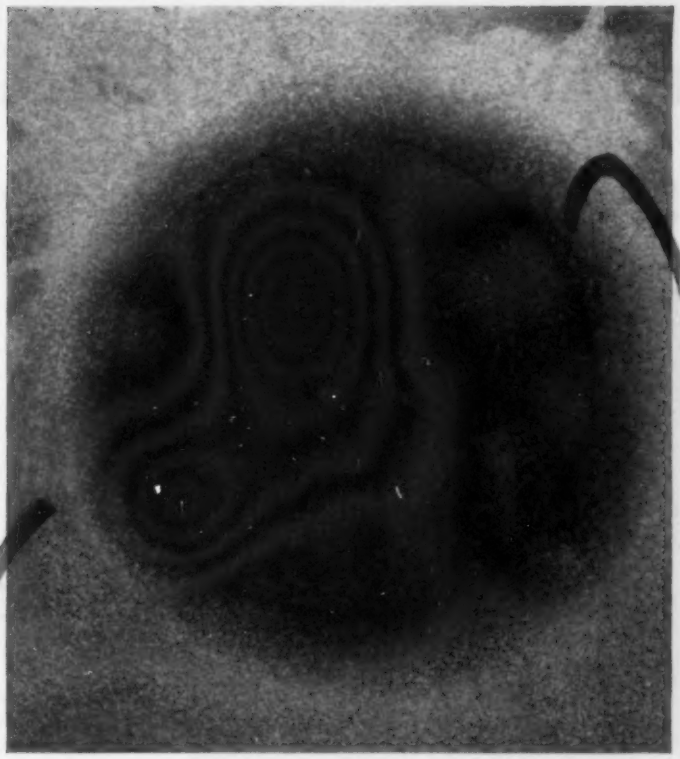

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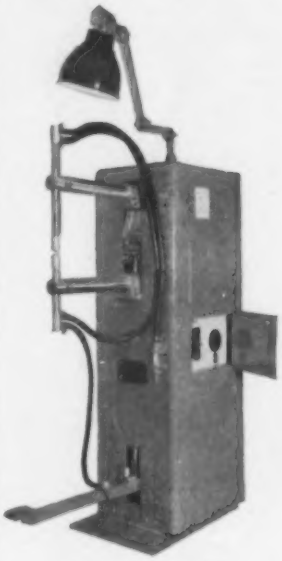

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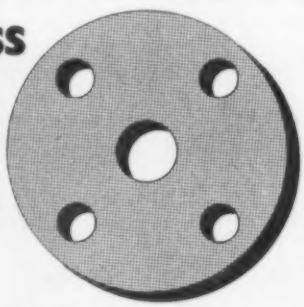

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A Rational Design Procedure for Pressure Vessels

To develop the designs of pressure vessels best fitted for specific purposes, it is necessary to consider each contingency to which the vessel is likely to be subjected in its life and to take account of each one separately in the design. The following paper sets out the likely contingencies and the material properties involved in designing for them. Commission XI of IIW intends to use this paper as a pattern for its future study of design problems, and it has taken steps to produce eight documents of up-to-date information, each one dealing with one of the contingencies mentioned.

EVER since engineers first began to design pressure vessels for industrial use, it has been their custom to calculate the thickness and shape of the vessel by a method which aims primarily at limiting the mean stress to an agreed allowable value, depending upon the steel to be used. In most countries the agreed values, which have been called "design stresses", are obtained by taking the minimum of agreed proportions of several of the physical properties of the steel, such as x% of the ultimate tensile strength, y% of the proof stress, and z% of the creep strength. In pressure vessel codes emphasis has, as a rule, been placed on designing so as to limit the stress to some single allowable value dependent only on temperature.

As it is intended in the future that pressure vessels should be designed closer and closer to the limits of the properties of the steels used, it is necessary that the designer should approach his problem in a new way, and he cannot expect to design his vessel from one tabulated allowable stress. He must acknowledge that there are a number of recognizable ways in which a pressure vessel may fail to function satisfactorily. For the design of some important vessels it may be necessary to take account separately, as suggested below, of each of these contingencies, almost every one of which is related to a different basic physical property of the material. To do this, data on the physical properties, shown in bold type in the following, must be used. It is the province of the designer to decide in advance which of these contingencies must be considered in any particular case.

EIGHT FACTORS TO BE CONSIDERED IN DESIGN

(1) Excessive Plastic Deformation

In regions where the stress is nearly uniform, both through the thickness and over the surface area, the maximum allowable stress* is theoretically the **proof stress** or **appropriate creep stress.** In regions where the stress is non-uniform, the maximum allowable stress will depend on the **proof stress** and the **ductility** or **appropriate creep data**, including **hot rupture data** and **relaxation.**

Document IIW/IIS-9-58 (ex Doc. XI-21-58) of Commission XI—*Pressure Vessels, Boilers and Pipelines* of the International Institute of Welding.

* Wherever a pressure vessel stress or strain is referred to, it implies the equivalent uni-axial stress or strain.

Two cases can be considered in relation to the type of stress distribution:

(i) If the stress system is such that yielding or creep at the maximum stress point causes a more nearly uniform stress distribution, as happens at a point of highly concentrated stress surrounded by elastic material or surrounded by a material with a relatively small creep rate, the peak theoretical stress range should be limited in such a way that it causes neither severe deformation nor loss of ductility.

(ii) If the stress concentration is of a more wide-spread character, in which yielding at the maximum stress point would lead to plastic instability or when creep would lead to shape instability, the maximum theoretical stress should be limited to that which would cause full plasticity over the cross-section under consideration, or to that shown in appropriate creep data, including relaxation.

(2) High Strain Fatigue

High strain fatigue is confined to stress concentration areas surrounded by less-stressed material capable of forcing the more highly stressed material to extend or compress. The maximum allowable strain range will be equal to the **fatigue strain range necessary to cause failure in N cycles**, depending on the duty of the vessel. It should be remembered that thermal as well as loading strains must be accounted for.

(3) Corrosion Fatigue

The procedure for designing to take account of corrosion fatigue is the same in principle as for high strain fatigue. The maximum allowable strain range occurring on any surface must not exceed a value determined from **corrosion fatigue data**.

(4) Stress Corrosion

The maximum allowable absolute stress occurring on any surface must not exceed a value determined from **stress corrosion data**.

(5) Metal Wastage by Corrosion and Erosion

A corrosion allowance is added to the calculated thickness, this allowance being based on **appropriate corrosion and erosion data**.

(6) Brittle Fracture

It is now fairly widely accepted that three conditions are required to cause a brittle fracture:

(i) The vessel temperature must be below the appropriate transition temperature of the material

(ii) There must be some defect, notch, or other such adverse shape to act as a crack-starter

(iii) The stress levels in the area must be sufficiently high to initiate and propagate the crack.

At least one of these conditions must be unfulfilled to prevent brittle fracture. Transition temperature is at present measured by several different methods giving differing results. It is the province of the designer to use the temperature obtained from the most appropriate form of test. The design and construction of the vessel should be such as to ensure that the material is able to yield to the point of plastic instability before it ruptures. It is almost impossible to assess whether this requirement is met in any particular instance where the residual stress distribution in the area is unknown and the possibility of an undetected flaw in construction cannot be eliminated. It is equally difficult to forecast the applied stress levels which will be safe where the residual stress levels are unknown. Therefore it is obviously desirable to select a steel whose appropriate **transition temperature** is below the minimum temperature at which the vessel will be loaded.

Where the residual stress distribution is known; for instance, following a furnace stress-relieving treatment, a margin of safety for rupture beyond the point of plastic instability may be predictable, even in the presence of an undetected flaw in construction, and where the minimum temperature at which the vessel is to be loaded is below the transition temperature.

(7) Excessive Elastic Deformation

The elastic strains occurring in the material adjacent to unwelded mechanical joints must be determined to ensure that the joint will remain tight. This procedure is usually carried out as a check on one or more of the design procedures given under items (1), (2), and (3). The basic material properties concerned are the **modulus of elasticity and Poisson's ratio**.

(8) Buckling

The stability of the vessel must be examined. A number of the physical properties of the material are involved such as **modulus of elasticity, Poisson's ratio, proof stress, and creep data**.

A feature of this approach to pressure vessel design is that a distinction is drawn between the use of absolute stress and stress range; also between absolute strain and strain range.

It should be noted that no allowance is made in this document for design to take account of bursting, because either excessive deformation, as described under (1), or brittle fracture, as described under (6), would inevitably occur before bursting; these paragraphs therefore cover this contingency.

Moreover, it should be noted that the design procedure given under sections (1), (6), (7), and (8) must be carried out for the test conditions as well as for the working conditions.

B.W.R.A. REPORT

Effect on Fatigue Strength of Internal Defects in Welded Joints

A SURVEY OF THE LITERATURE

By R. P. Newman,
A.M.I.MECH.E.

It is sometimes assumed that the effect of weld defects on strength has received no attention from research workers. This is not correct, though the overall problem is complex, and difficulties of experimental technique are severe. A moderate amount of useful work has nonetheless been achieved, and that part of it dealing with fatigue properties in particular is reviewed in this paper.

The survey does not claim to be complete, but it examines the present knowledge of the influence of defects in butt welds in mild steel plate, since this is considered to be the most important field for immediate research.

THE effect on the strength of welded joints of those internal defects normally revealed by non-destructive methods of examination commands wide interest. This interest is often associated with the possibility of establishing a code of practice of acceptance limits for internal defects. Support for such codes, however, is by no means unanimous, for the need to maintain personal judgment in the assessment of minimum quality is often expressed. Whatever the respective merits of these two viewpoints, and it is not intended to discuss them here, there is undoubted general support for the need to continue, and greatly expand, research into the influence of weld defects.

Much of the early experimental work on the properties of defective welds was related to the influence of defects on static tensile strength and these studies will therefore not be dealt with here. It may be mentioned, however, that they have repeatedly confirmed the general experience of static tensile testing of fusion welded joints in mild steel that quite high levels of defect severity have little, if any, influence on joint strength. A recent demonstration of this is provided by the work of Green, Hamad and McCauley[1] who evaluated the influence of porosity on the mechanical properties of mild steel weld metal. Tensile, bend, and impact tests were carried out, and in each case the related property was not influenced by porosity up to a severity level of 7% (measured as a percentage reduction in cross-sectional area). In the impact tests, Green and his co-authors used standard Charpy V-notch specimens, with the notch located in the side face of a butt weld so that the weld metal was fractured.

In this way a direct measure of the properties of the defective deposited metal was obtained. A different approach was adopted in some exploratory impact-tension tests conducted for the British Welding Research Association by the National Physical Laboratory. These tests have not yet been reported, and a brief description of them here may not be out of place.

Impact-Tension Tests

The exploratory tests were carried out to establish whether, on transverse, metal-arc butt welds at room temperature, the loading condition of high energy impact-tension would prove to be more sensitive to the presence of defects than would static tension loading. Three main groups of specimens, as shown in Fig. 1, were constructed from ½ in. thick mild steel to BS. 15:

(i) Plain plates (3 specimens)
(ii) Good quality transverse butt welds (4 specimens)
(iii) Defective transverse butt welds (8 specimens)

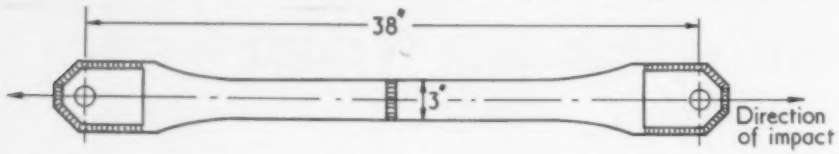
The test pieces were broken under impact-tension in the large N.P.L. machine, fracture being produced usually with the second blow. Each blow had an initial energy of 12 ft tons. In the welded specimens the fractures were located in the parent plate away from

Report D2/2/58 of the British Welding Research Association. Mr. Newman is Chief Research Engineer to the Association.

the joint, so that no significant differences in energy to failure were observed either between groups (ii) and (iii) or between the welded specimens generally and the plain plates. The energy-to-failure values recorded were within the range 16–22 ft tons.

Two of the good quality welds in group (ii) were made with class E.317 electrodes, and two were made with E.615 electrodes. As with group (iii) specimens, the weld reinforcement was left on and no stress-relieving was applied. Before testing, every welded joint was examined radiographically and the term 'good quality' as used here refers to freedom from anything but quite minor defects in the X-ray film.

The intentional defects included in the group (iii) specimens are listed in Table I. Radiographs taken after testing indicated that only the longer cracks had been propagated, and their extension was slight. Thus the general behaviour of the defective specimens under impact-tension loading must be judged to be the same as under static tension.



1—*Impact-tension specimen in ½ in. thick mild steel to BS.15*

Previous Work on the Influence of Defects on Fatigue Strength

The five main types of defect discussed are: cracking, lack of fusion, lack of penetration, porosity, and slag inclusions.

Cracking

Possibly the most successful investigation of the influence of weld metal cracking, in the experimental sense, was that carried out by Warren,[2] whose results support the contention that cracks are damaging to fatigue strength, especially at high endurance values. His specimens were transverse, manually welded butt joints, $2\frac{3}{8}$ in. wide by $\frac{1}{2}$ in. thick, containing fine internal cracks lying parallel to the weld direction. The cracks were produced by depositing a small root run under conditions of high restraint. The limits within which the crack area could be controlled are not known but an illustrated example of a fracture surface indicates a crack equivalent to about 5% of the total cross-sectional area. An alternating load cycle was used, and sufficient experimental points were obtained to plot a band of results on an S/N diagram (Fig. 2) extending to ten million cycles, at which the upper and lower limits of the band correspond to stress levels on the plate of approx. $\pm$ 2·25 and $\pm$ 3 tons/sq.in. respectively. Taking the average stress of $\pm$ 2·6 tons/sq.in., a 60% reduction of strength could thus be assumed as a rough estimate of the influence of the cracking, based on strengths at 10 million cycles. This estimate would have more value if it were known that the 5% severity of defect had been maintained reasonably constant.

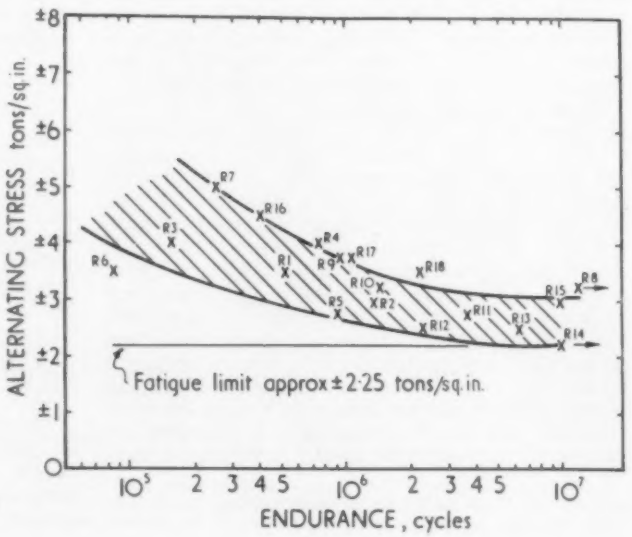
Lack of fusion

For this defect also, there exists only a small amount of information on its effect on fatigue strength. This was obtained from the B.W.R.A. investigation on pipe butt joints[3] in which one series of specimens was made containing extensive lack of fusion. The pipe specimens were tested under alternating bending, with the welded joints normal to the direction of applied stress; the plane of the defect circumferentially was also normal to the direction of stress, but through the thickness of the weld it followed the original 40° pipe

Table I

Defective specimens tested under impact-tension

Specimen	*Electrode Classification*	*Intentional Defect*
1 2	E. 317 E. 615	Crack ⅜ in. long, ½ in. deep, longitudinal with respect to weld
3 4	E. 317 E. 615	Crack 1 in. long, ½ in. deep, longitudinal with respect to weld
5 6	E. 317 E. 615	Slag line, 1 in. long
7 8	E. 317 E. 615	Slag line, 2 in. long



2—*Fatigue strength of manually welded butt joints with fine crack* (Warren[2])

edge preparation from the outside surface to a point at about mid-thickness of the pipe wall. In the tests on these specimens, the fatigue cracks were consistently initiated from the root area of the weld at the transverse discontinuity formed by the junction of the backing ring with the pipe. A fairly thorough examination after testing revealed that only in one instance had a fatigue crack been initiated from the tip of the defect, and that this crack was very small and localized.

In making these specimens it was very difficult to produce lack of fusion with metal-arc welding, and the defect was eventually produced by the use of oxy-acetylene welding. The frequency of occurrence of lack of fusion in metal-arc welded construction is something which might, therefore, be considered carefully before making any proposals for investigating this fault.

Lack of penetration

As used here, the term 'lack of penetration' refers to a condition in butt welds, either local or general in extent, where the thickness of the weld metal contained within the plate surfaces is less than the thickness of the plates joined. Some joints may be so designed that full thickness welds are not obtained (*e.g.*, nozzle connections in pressure vessels); these are usually described as 'partial penetration' joints. Unwelded zones thus become defects only when full thickness welds are necessary.

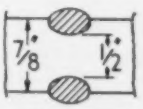
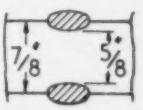
As a defect, lack of penetration can obviously occur, in terms of the area criterion, at almost any degree of severity, and it can also occur either as a 'buried' defect, where plates are welded from both sides, or as a 'surface' defect where welding from one side is involved. In flat plate specimens containing transverse welds, the buried and surface forms might be expected to give different results because of the eccentricity of loading, but no experimental observations are available. Tests on the surface type were carried out with pipe specimens in the B.W.R.A. investigation[3] and, although quite positive results were obtained, eccentricity effects would not have operated in the circumferential joints used. These tests were designed to show the influence of *continuous* lack of penetration, $\frac{1}{8}$ in. deep, at the root of butt welds in $\frac{3}{4}$ in. thick pipe. In this severe form the defect produced a marked loss of fatigue strength, measurable as a 60% reduction of stress range at 2 million cycles.

Two other investigations also have dealt with the influence on fatigue strength of continuous lack of penetration, and these refer to butt joints in flat plate welded from both sides of the joint. Tests on transverse butt welds made by the submerged-arc process have been reported by Warren.[2] No data are given for sound welds and the severity of the defect can again be judged only from an illustration of one of the fracture surfaces. This makes it clear that the lack of penetration existed at mid-thickness of the weld and was continuous over the full width of $2\frac{3}{8}$ in.; the defect varies in depth, but the average appears to be $\frac{3}{32}$ in., that is, about 20% severity on an area basis. For these welds Warren quotes a fatigue strength (at 2×10^6 cycles) of approx. $\pm 3\frac{1}{2}$ tons/sq.in. on the plate area. Without control data it is difficult to judge the reduction of strength brought about by the defect, but a 50% reduction would probably be a maximum for transverse submerged-arc welds with the reinforcement left on.

The specimens used in the second investigation, carried out by Wilson, Munse and Snyder,[4] are defined as partial penetration butt welds, since the unwelded areas were intentionally kept high (40–60%). The tests were carried out on manually welded specimens in $\frac{7}{8}$ in. thick *A*7 mild steel plate, and both transverse and longitudinal joints were studied. For the longitudinal joints it was shown that the presence of the large unwelded areas had no effect on the fatigue strength, the discontinuity being oriented in the same direction as the applied load. For the transverse joints, severe reductions of strength were obtained, as shown by the summarized results in Table II. There is little difference between the 40% severity specimens with or without the weld reinforcement, whereas at 60% severity there is a small difference. There was also a slight general increase in the strength of the lower severity specimens as compared with those of 60% severity. This suggests a closer correspondence between weld area stresses, and it is found that the range is 4·8 to 6·2 tons/sq.in. (*i.e.*, 23% variation as against 36% variation in plate stress).

Table II

Fatigue strengths for transverse partial penetration butt welds[4]

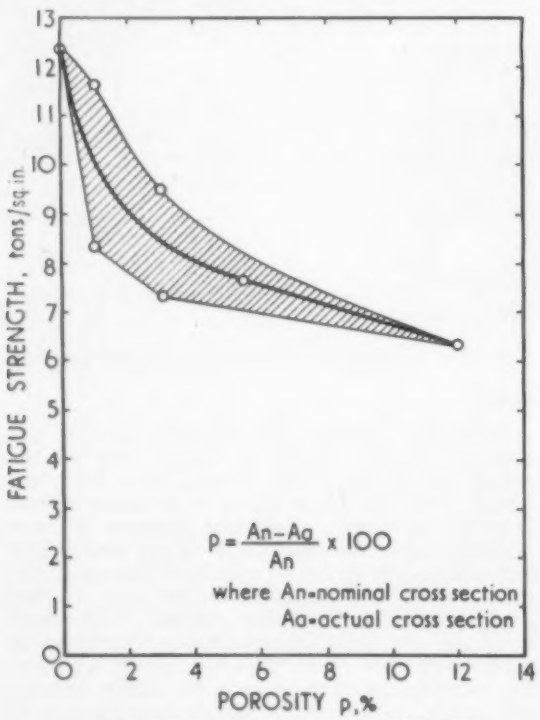
Specimen	*Weld Reinforcement*	*Fatigue Strength (pulsating tension) on plate area at 2×10^6, tons/sq.in.*	*Reduction of Strength % (10 tons/sq.in. = control value)*
 40% severity	On	3·6	64
	Off	3·75	62
 60% severity	On	3·1	69
	Off	2·4	76

Porosity

In codes of practice the tolerance on this very common defect is specified in several ways, but the acceptance of some degree of porosity is quite universal. This may be based on a comparison with standard radiographs (American practice) or on a comparison with the radiographs of test welds which have successfully passed the required mechanical tests (British practice); in certain codes the limits of tolerance are defined in distribution diagrams. Each system recognizes the difficulty of defining the defect in quantitative terms—a problem that equally applies to any experimental work where a correlation between the severity of the defect and fatigue strength is attempted.

One of the best known investigations of the influence of porosity on fatigue strength was carried out by Homès[5] who used the photometric method to assess the severity of the defect. By this method, measurements of X-ray film density are made and compared with measurements from a uniform standard to derive an effective or actual cross-sectional area for the zone concerned. Thus defect severity, in quantitative terms, is expressed by the reduction in area caused by the porosity (in line with the criterion referred to in previous sections). The dependence on density measurement appears to limit the use of the method to machined specimens and, even then, the assumed linearity between effective specimen thickness and film density measurement can be questioned.[6] However, as far as is known, it is the only method available for arriving at a quantitative assessment of the defects shown in radiographs.

The specimens used by Homès were mild steel plates, $\frac{3}{4}$ in. thick by $1\frac{1}{2}$ in. wide, containing transverse butt welds with differing amounts of porosity. The specimens were fully machined before testing and radiographic examination. The pulsating tension endurance limits were determined for the test pieces, and the results were plotted against the percentage of porosity, as in Fig. 3. With no porosity a fatigue strength of $12\frac{1}{2}$ tons/sq.in. was obtained. There was then a rapid decrease in strength up to 2–3% porosity, after which the decrease was less severe up to a maximum of 12% porosity. The scatter of results appears to be greater for the smaller amounts of porosity; for example, at the 1% level the strength limits are approx. $8\frac{1}{4}$ and $11\frac{3}{4}$ tons/sq.in. The upper limits in this area of the diagram suggest that there would be virtually no influence on the fatigue strength of a transverse butt joint for porosity levels up to 3% if the weld reinforcement were left on. The same would not apply at the lower limits, but in interpreting the results in this way it should be recognized that the tests were reported in 1938 and improvements in the properties of weld metal since that time might well modify the tolerance limit for porosity.



3—*Effect of porosity on fatigue strength* (Homès[5])

A more recent investigation of the possible correlation between fatigue strength and, in this instance, porosity and slag inclusions, has been carried out by Masi and Erra.[7] They also used the photometric method to determine defect severity in fully machined specimens, and established five classes of specimen, numbering about twenty in each class. The step method of fatigue testing was employed, and a fixed mean stress of $9\frac{1}{2}$ tons/sq.in. was used throughout; endurance limits are stated as stress ranges for which the mean value is, therefore, constant at $9\frac{1}{2}$ tons/sq.in. Table III outlines the results obtained, as applied to both porosity and slag inclusions.

Table III

Relationship between defect severity and fatigue strength[7]

Class	*Defect Severity, %*	*Mean Endurance Limit and Scatter, tons/sq.in.*
0	0	17·3 (± 1·4)
1	0·8	14·3 (± 1·17)
2	4·37	10·7 (± 0·88)
3	9·36	7·1 (± 0·82)
4	14·4	5·5 (± 2)

It seems reasonable to suggest, however, that had the specimens contained porosity alone, the fatigue strength values in classes 1–4 would not have been reduced; a direct comparison with Homès' results might in fact give a conservative estimate of the improvement of the strength of defective weld metal that seems to be evident between the two investigations. This comparison is made in Table IV.

Table IV

Comparison of results obtained by Homès[5] and Masi and Erra[7]

Defect Severity %	*Fatigue Strength Range*			
	Homès		*Masi and Erra*	
	Upper limit	*Lower limit*	*Upper limit*	*Lower limit*
0	12·5	12·5	18·7	15·9
1	11·75	8·25	15·47	13·13
4	8·7	7·25	11·58	9·82
9	7·2	6·6	7·92	6·28

Up to a severity of 4% the later results of Masi and Erra are clearly superior to those of Homès. Basic electrodes were used by Masi and Erra, and it appears possible that with such electrodes a value of 4% severity for porosity could be tolerated as an upper limit in transverse butt welds with the reinforcement left on, at least from the standpoint of fatigue strength. At 9% severity the stress ranges obtained from the two investigations are quite similar, but whereas in Homès' tests the mean stress is always equal to half the range, in the Masi and Erra tests it is greater than

the range. This variation in mean stress could account for the non-constant difference in stress ranges between the two sets of results.

The possibility that porosity levels up to 3 or 4% would not influence the fatigue strength of a butt weld with the reinforcement left on can be checked only to a very limited degree from the current B.W.R.A. programme, in which the incidence of porosity in manual butt welds has been observed in radiographs. Some difficulty arises in expressing defect severity for these specimens, but for $\frac{1}{2}$ in., and $1\frac{1}{4}$ in. thick welds made with normal penetration electrodes a rough estimate indicates the worst conditions to be:

(*a*) $\frac{1}{2}$ in. welds—five pores of $\frac{1}{32}$ in. dia. per inch on one plate

(*b*) $1\frac{1}{4}$ in. welds—ten pores of $\frac{1}{64}$ in. dia. per inch on one plate

Both conditions can be graded as less than 1% severe. It should be mentioned however that in the $\frac{1}{2}$ in. welds the defect occurred substantially as a single line of pores in the root area, whereas in the $1\frac{1}{4}$ in. welds the porosity was uniformly scattered across the width of the radiograph image. The criterion of severity adopted makes no distinction between aligned and scattered porosity and would rate the two conditions equal if the reduction in area was the same *on a given plane*. This point may be worthy of experimental study, but for the present it can be noted that in neither instance did the defects influence the fatigue behaviour of the transverse butt weld specimens, for failure occurred consistently from the edge of the reinforcement.

Greater defect severity was obtained in $\frac{1}{2}$ in. butt welds made with deep penetration electrodes, where the pore size varied over wider limits (up to approx. $\frac{3}{32}$ in.) and severities approaching $1\frac{1}{2}$ or 2% were probably obtained. In this instance, however, fatigue behaviour can be less well defined, since other factors such as bad reinforcement shape and surface decarburization have to be considered. The combined influence of these two factors was more pronounced than that of the porosity, and it cannot be definitely concluded that the same degree of porosity would prove innocuous in otherwise good welds. Another aspect of the problem concerns the possible difference in behaviour between transverse and longitudinal butt welds. Earlier, defect severity has been related to transverse welds only and has been estimated by reference to the reduction in load-transmitting area on a given plane or section normal to the direction of applied stress. It is clear, however, that if another plane at right angles to the first is considered, then for the same porosity condition, a different degree of severity might be obtained, particularly with aligned porosity. It is possible therefore that a defect could give different results according to whether it occurred in transverse or longitudinal welds.

Hempel and Möller[8] carried out tests wherein such an effect was observed. Their tests were made on *ST*37 mild steel specimens classified, according to radiographic appearance, as:

(i) Without defects
(ii) With isolated pores
(iii) With many pores
(iv) With a very great number or pores

Fully machined longitudinal and transverse joints gave identical results for categories (i), (ii), and (iv), but in category (iii) the longitudinal joints were superior, giving a fatigue strength of $11\frac{1}{2}$ tons/sq.in. as compared with 7–8 tons/sq.in. for transverse joints. The authors give the following explanation of these results:

> "In the case of specimens with isolated pores (ii) a given cross-section of the specimen will normally contain one single pore only, no matter whether the load acts normal or parallel to the weld. The notch effect will therefore cause the incipient fatigue fracture, independent of the position of the weld. With specimens containing many pores (iii), however, the possibility of several pores occurring in the same cross-section is considerably greater for specimens with transverse welds than for specimens with longitudinal welds. For that reason, the incipient fractures starting at the pores will, in the case of transverse specimens, combine comparatively soon and thus bring about an early fatigue failure.
>
> "With longitudinal specimens, however, the reduction in the cross-section will occur much more slowly even in the case of specimens containing many pores, because the pores will normally be present in different cross-sections at different levels so that a given cross-section is likely to develop a single fracture only. But if the specimen contains a very great number of pores (iv), the difference will again disappear because, independent of the position of the weld, each cross-section is likely to contain several incipient fatigue fractures simultaneously so that the weld will fail within approximately the same time, whether it be a transverse or longitudinal weld".

For the longitudinal joints Hempel and Möller used specimens 2 in. wide and they recognized that the residual cross-section of parent plate would not have delayed final rupture to any great extent. The strength

Table V

Results of tests on welded specimens with differing amounts of porosity[8]

Type of Specimen	*Condition*	*Defects*	*Fatigue Strength (pulsating tension), tons/sq.in.*
*ST*37 plain plate	Machined	—	22·8
	As rolled	—	19
Transverse butt Joint	Machined	Sound	14·6–19
		Isolated pores	15 –18
		Many pores	6·9 –8
		Great number of pores	4·4 –6·3
	As welded	Sound	13·3
		Isolated pores	12·6
		Many pores	8·2
		Great number of pores	6·3
Longitudinal Butt Joint	Machined	Sound and isolated pores	15
		Many pores	11·4
		Great number of pores	5·7

of longitudinal specimens at rupture could be influenced by specimen width to a greater degree than for transverse joints, but experimental proof would necessarily be restricted by testing machine capacity.

It is not possible to state the results of the tests by Hempel and Möller in relation to defect severity, other than by using the arbitrary grading given in Table V. The results are of interest, however, and are summarized in the Table. In all cases isolated pores had no effect on fatigue strength.

As far as can be judged from the radiographs reproduced in the paper, defect severity for 'isolated pores' was less than that associated with the B.W.R.A. specimens previously referred to, whilst the condition 'many pores' was considerably more severe, and is perhaps the most interesting one to study. The difference between the machined, transverse, and longitudinal joints has already been described, but a further comparison can be made between as-welded transverse joints containing on the one hand 'no defects' and on the other 'many pores'. The sound joints have a high fatigue strength of 13·3 tons/sq.in., and there is a significant drop in strength to 8·2 tons/sq.in. for the defective joints. Thus, although the latter value is quite good in itself, a reduction of nearly 40% could be attributed to the presence of the porosity. But the authors state that:

> "... the fatigue fractures mostly take their origin at uneven points of the bead, especially at small undercuts, and less frequently at the pores."

This reduces the force of any argument about the influence of the porosity, since it is apparent that there was a marked difference in surface condition between the sound and the defective welds.

It remains to be considered why the *machined* transverse joints with 'many pores', from which any influence of external surface had been eliminated, did not have a higher strength than the as-welded transverse joints (6·9–8 tons/sq.in. as compared with 8·2 tons/sq.in.). Firstly, the authors have pointed out that in the latter specimens the number of pores was ... 'slightly but consistently' ... smaller than the number of pores in the machined specimens. Furthermore, the scatter in results for the machined specimens was somewhat greater, and it seems possible on the basis of fracture appearance that fatigue cracks had sometimes been initiated from defects exposed on the surface by machining.

Slag inclusions

Unlike porosity, no assumptions can be made about the general shape of slag inclusions, and since accurate sectional measurements would be difficult to make in non-destructive examination, the determination of severity appears to be limited at best, to a check of plane dimensions. The nature of the defect is often described by reference to its form; thus there are 'isolated inclusions' of which 'slag patches' may be a particular example, and 'elongated inclusions' of which 'tram-line' (or 'parallel line') inclusions are common instances. Something of the difficulty of using these terms, or even the dimensional basis, for assessing defect severity is illustrated by Warren's paper[2] which gives the results of three tests on transverse butt joints each containing a 'heavy slag patch'. At ± 8 tons/sq.in. the first two specimens failed through the weld after 62,800 and 85,000 cycles respectively; at the same stress the third specimen sustained 800,000 cycles and then failed away from the weld.

Warren obtained equally wide scatter limits on the number of cycles for nine specimens containing a 'heavy slag line at root'. Again tested at ± 8 tons/sq.in., these specimens had lives ranging between 21,500 and one million cycles. Better consistency was obtained from a group of specimens containing less severe slag lines, in different forms; the individual results for a stress range of ± 8 tons/sq.in., were:

Defect	*Life, cycles*
Faint intermittent slag lines	152,000
" " " "	526,000
Intermittent parallel slag lines	121,000
Parallel slag lines	347,500
Heavy parallel slag lines (one specimen)	49,000

The interesting feature of Warren's tests is that the apparent severity of the defects studied, as judged by the radiographs, is not grossly variable within each classification. The results are therefore rather surprising—especially for the specimens with heavy slag lines at the root. In an earlier investigation, Hempel[9] tested *ST*37 mild steel specimens containing equally severe inclusion defects and claimed good S/N relationships for groups of specimens also classified according to radiographic appearance. In a number of instances, however, he shows two S/N curves for a given group, stating that this is justified on the grounds that two degrees of radiographic severity can be identified within the group concerned. It is difficult to accept this from the evidence of the reproduced radiographs, and it is equally difficult to accept Hempel's conclusion that one can distinguish six degrees of severity for the 'slag inclusion' type of defect, each with its own distinctive fatigue strength value somewhere between 3·8 and 8·0 tons/sq.in. These difficulties may arise because of the considerable problem of reproducing radiographs without loss of quality, but if one discounts this critical judgment of the radiographs the scatter in life obtained by Hempel is of the same order as that reported by Warren.

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Welding
RESEARCH

Presentation of Fatigue Test Results

By J. Strebelle

It is shown how simple statistical methods can be used in the presentation of the results of fatigue tests. Three ranges of stress cycles are considered. An example is given of graphical representation on a Wöhler diagram.

ANY set of measurements, however carefully made, has some scatter of the results around the mean value. If the scatter is wide it is not easy to obtain information from the data, and it is even doubtful whether the mean value has any significance. This is particularly so for fatigue tests, since their results are naturally scattered.

In a set of correctly performed tests, the results are not arbitrarily grouped around the mean value, and deviations from it therefore become rarer as they increase in magnitude. From this observation have emerged statistical rules for interpreting the results of fatigue and other tests.

Most of the information obtainable from an experiment is lost if only the mean of the results is considered. The mean must always be supplemented by other information, as will be explained later. In a set of measurements, any systematic error should be carefully eliminated because statistics deal only with casual, random deviations.

Within each 'sample' of test pieces, there should be reasonable certainty that they can be considered to belong to the same 'population', having the characteristics which are to be studied, and that they are unaffected by outside influences. Document 13–100–55 of the French delegation can be summarized as follows: "Statistics of fatigue tests must be based on correct assumptions confirmed by numerous observations." Also "The information obtainable from statistics is proportionate to the amount given by the engineer to the statistician."

Translation of document 13–153–58 of the Belgian delegation to the International Institute of Welding, based on documents 13–53–54 and 13–100–55 with the help of the French delegation. Originally published in *Revue de Soudure*, 1958, vol. 14, pp. 24–32.

Three different stress zones must be considered in the statistical study of fatigue tests:

(i) High stresses
(ii) Stresses at or near the fatigue limit
(iii) Stresses slightly above the fatigue limit.

Large numbers of fatigue tests have shown this distinction to be necessary. Because the results in these three stress zones have a different structure, different statistical treatments are needed. In this article, a stress cycle is considered to vary from 0 to a tensile stress T.

Zone (i) Tests at High Stress

The distribution of results (number of cycles to rupture for a given amplitude) is fairly well represented by a log-normal (log N) distribution. However, theoretical distributions even when they fit the experimental results closely, should not be extrapolated far outside the area where their validity has been checked. In the present instance the risk of error in applying the log-normal distribution increases at the ends of the distribution.

From the practical viewpoint of this article and because of the absence of wholly reliable information on the distribution of results towards the ends, the log-normal distribution must unavoidably be adopted. But the risk of errors is negligible within the range 5–95%, becoming appreciable outside this interval, and large outside the 1–99% interval. All that needs to be done therefore, is to look up the logarithm of each result, and to consider these logarithms as forming a normal distribution.

Unbiased estimates of the mean value of the

Table I
Correction factor and Student's $t(n, \alpha)$

	Number of Specimens, n										
b_n	4 0·798	6 0·869	8 0·903	10 0·923	15 0·949	20 0·962	30 0·975	50 0·985	75 0·990	100 0·992	∞ —
α											
90%	2·35	2·01	1·89	1·83	1·76	1·73	1·70				1·64
95%	3·18	2·57	2·36	2·26	2·14	2·09	2·04				1·96
99%	5·84	4·03	3·50	3·25	2·98	2·86	2·76				2·58
99·9%	12·94	—	5·96	—	—	—	—				3·09

logarithms, the variance, and the standard deviation can be obtained as follows:

$$\text{Mean value } \overline{\log N} = \frac{\Sigma \log N_i}{n}$$

$$\text{Variance } S^2 = \frac{\Sigma (\log N_i - \overline{\log N})^2}{n-1}$$

$$\text{Standard deviation } \sigma'' = \frac{S}{b_n}\sqrt{\left(\frac{n-1}{n}\right)}$$

where n = number of test pieces

N_i = number of cycles to rupture of the i-th test piece

b_n = correction coefficient (Table I).

Values of Student's $t(n, \alpha)$ are also given in Table I; their use is explained later.

In any example, the following can be deduced from these formulae:

(*a*) An interval having 9 chances in 10 of containing the true mean value, $\overline{\log N}$ ($t(n, 90)$ being found from Table I at the intersection of the 90% value and n, the number of specimens tested):

$$\overline{\log N} = \log N \pm t(n,90)\,(S/\sqrt{n})$$

(*b*) The number of cycles N_1 for which one can expect $\frac{(100-90)}{2}$ per cent of unbroken specimens:

$$\log N_1 = \overline{\log N} + t(\infty,90)\,\sigma''$$

(*c*) The number of cycles N_2 for which one can expect $90 + \frac{100-90}{2}$ per cent of unbroken test pieces:

$$\log N_2 = \overline{\log N} - t(\infty,90)\,\sigma''$$

Even more information, beyond the scope of this article, can be obtained by a statistician from these results. It follows from the foregoing information that the results from a sample of fatigue specimens at high stress can be plotted on a Wöhler diagram (Fig. 1), the mean of the series being represented by a point with the coordinates ($\overline{\log N}$, T). This representation is completed by a horizontal straight line which marks the 90% confidence zone for the mean,

$$\log N \pm t(n,90)\,(S/\sqrt{n}),$$

and the 90% confidence zone for the individual results,

$$\overline{\log N} \pm t(\alpha,90)\sigma''$$

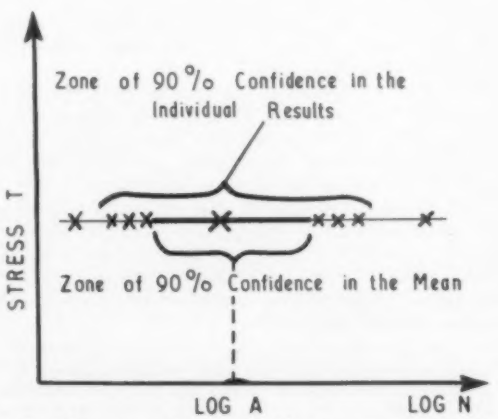
Results presented like this show clearly the degree of confidence which the reader can have in them, and enable him to make the best use of them, at least for results in the mean. When it is stated that, for so many cycles, 95% of survivors can be expected, what is meant is 95% average; thus there will be 93% in one sequence of experiments, 96% in another, and so on.

The question can be put in this way; what is the highest number of cycles in which one can be reasonably sure of 95% of survivors? One must bear in mind that σ'' is only an estimate of the standard deviation of the population and another value should be taken which is reasonably sure to be not less than the true σ (χ^2 test). This is one reason why a statistician should be consulted. These formulae show the importance of a large number of tests in raising the confidence level of the results.

Special case

It may happen that some specimens of a sample have not broken, merely because their testing time has been intentionally short. Fracture is not physically impossible, but it could occur if the test were prolonged. It is still possible to draw conclusions from such a sample, where some of the data are undetermined, but the statistical results are less accurate than those from a completely determined sample.

The following procedure must be carried out. The information shown in Table II is drawn up, putting the results in the order of increasing number of cycles to fracture, starting from the earliest number of cycles N_1. The results from columns 3 and 4 are plotted on



1—*Confidence zones plotted on Wöhler diagram. Each point indicates one test result*

probability paper with linear-Gaussian co-ordinates, as shown in Fig. 2.

Table II

Tabulation of Values

Index	*Number of Cycles to Rupture*	*Logarithm of these Numbers*	*Index divided by n+1*
1	N_1	$\log N_1$	$\frac{1}{(n+1)}$
2	N_2	$\log N_2$	$\frac{2}{(n+1)}$
3	N_3	$\log N_3$	$\frac{3}{(n+1)}$
i	N_i	$\log N_i$	$\frac{i}{(n+1)}$
$n-4$	$N_{(n-4)}$	$\log N_{(n-4)}$	$\frac{(n-4)}{(n+1)}$
$n-3$	$N_{(n-3)}$	$\log N_{(n-3)}$	$\frac{(n-3)}{(n+1)}$
$n-2$	$N_{(n-2)}$	$\log N_{(n-2)}$	$\frac{(n-2)}{(n+1)}$
$n-1$	undetermined	undetermined	$\frac{(n-1)}{(n+1)}$
n	undetermined	undetermined	$\frac{n}{(n+1)}$

A straight line is drawn through the points plotted. At abscissa 50%, an estimate of log N_{50}, the median, is read off the graph. Similarly at abscissa 99%, log N_{99} corresponds to the number of cycles above which 99% of broken specimens could be expected.

Half the difference between the ordinates at abscissae 16% and 84% gives an estimate σ of the standard deviation. From this a confidence interval for the theoretical mean can be obtained using the median of the results:

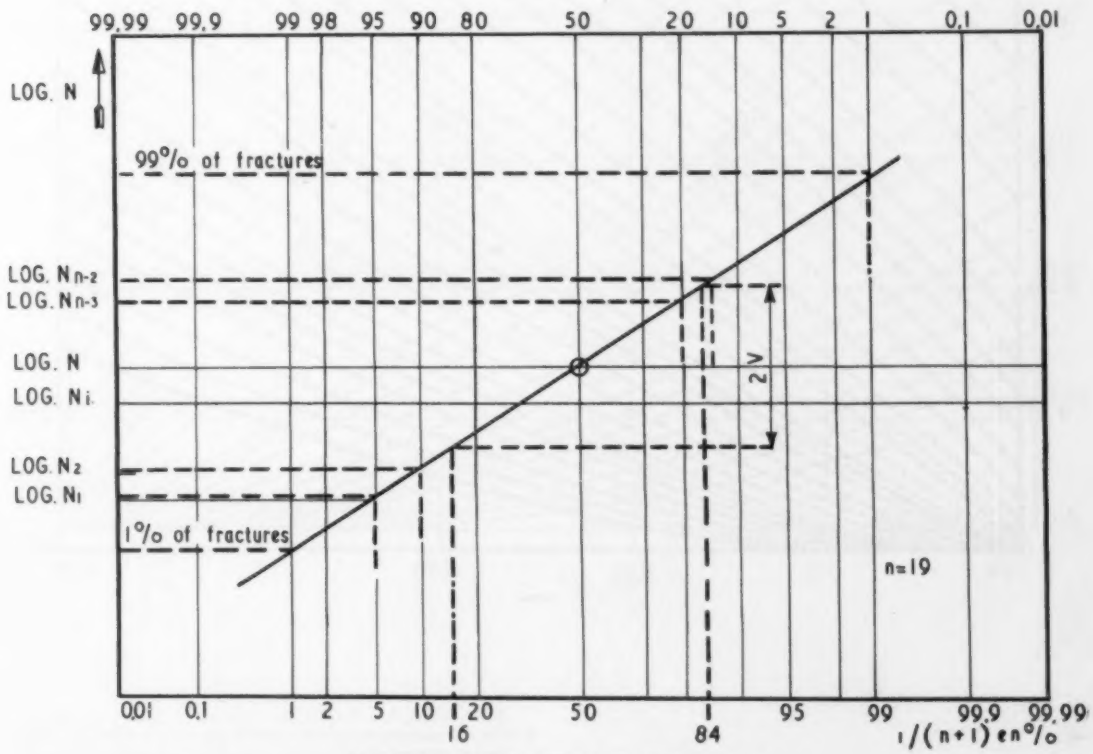
$$\text{median } (\log N_{50}) \pm t\,(\alpha,90)\,(\sigma/\sqrt{n})\,.\,1{\cdot}25$$

But this interval is not as exact as the confidence interval obtained by calculation from a complete sample, except for large values of n, say above 30. Quantities similar to those calculated previously can be obtained, but with less accuracy. This method is an improvement of that suggested in document 13–53–54 and is simpler than the method of the generalized range.*

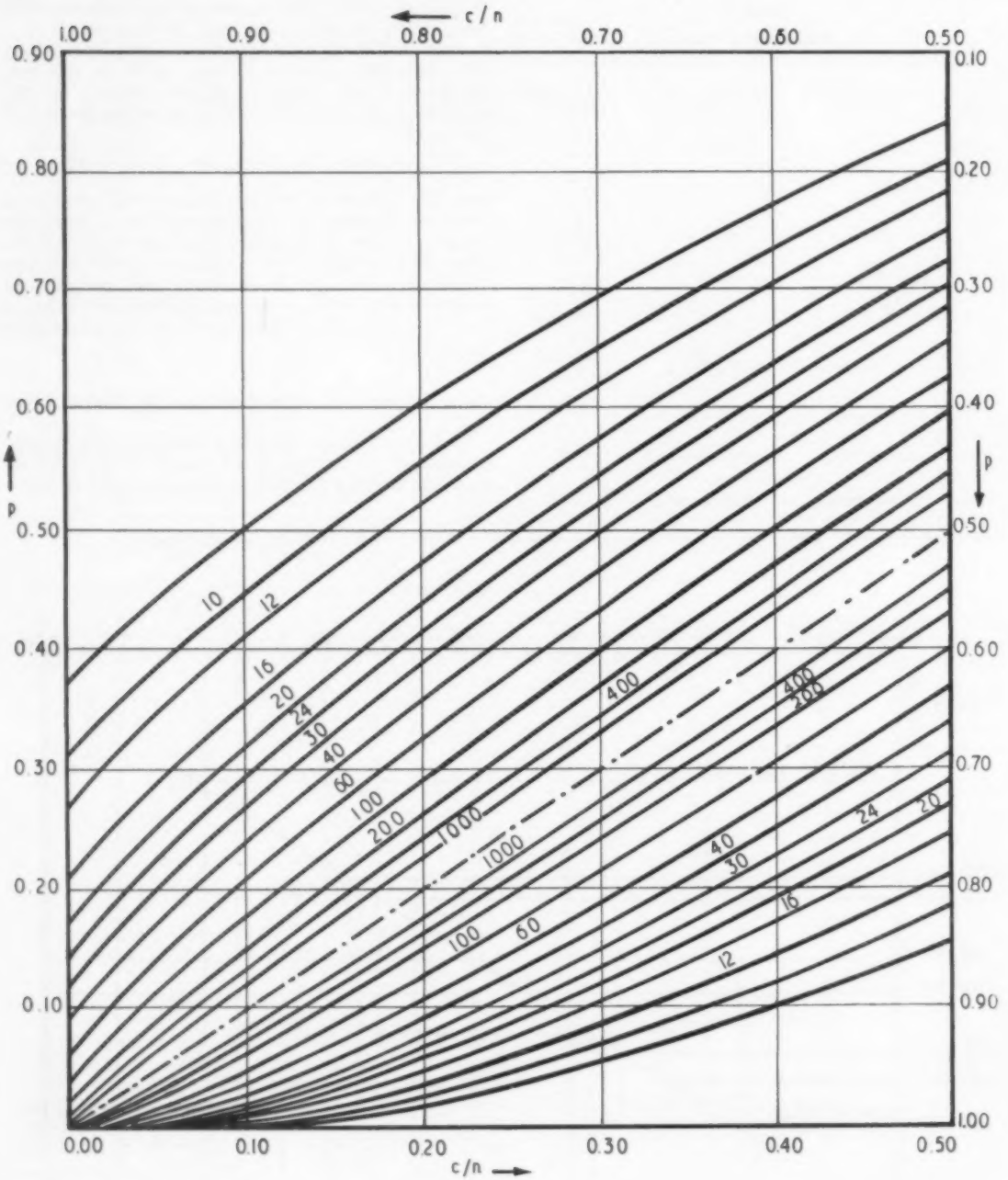
Zone (ii) Samples Stressed to Near the Fatigue Limit

In this zone there will obviously be an appreciable proportion of unbroken test pieces. Tests such as these are stopped at a predetermined number of stress

* Document 13–99–55 of the French delegation.

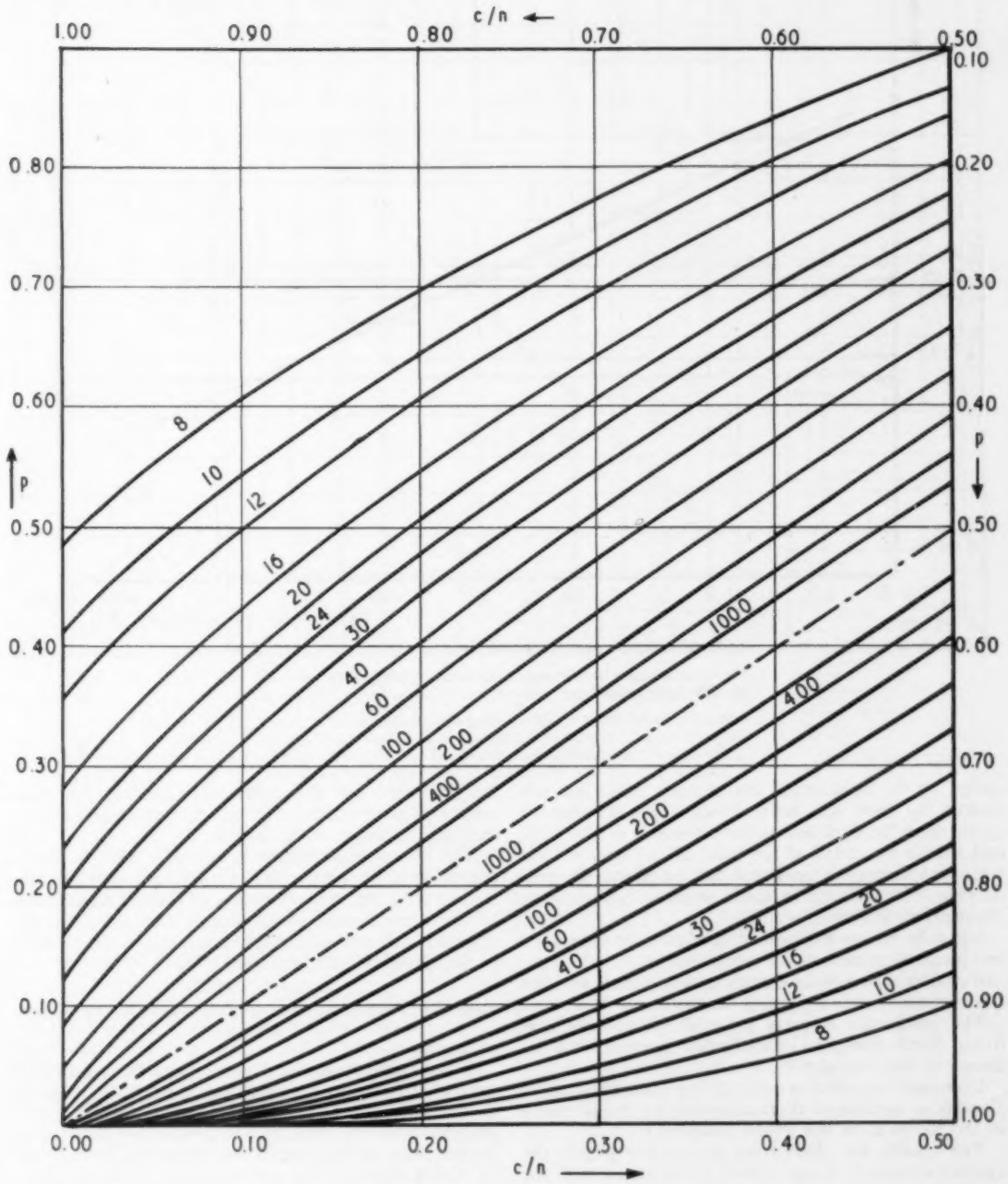


2—*Results plotted on linear Gaussian diagram*



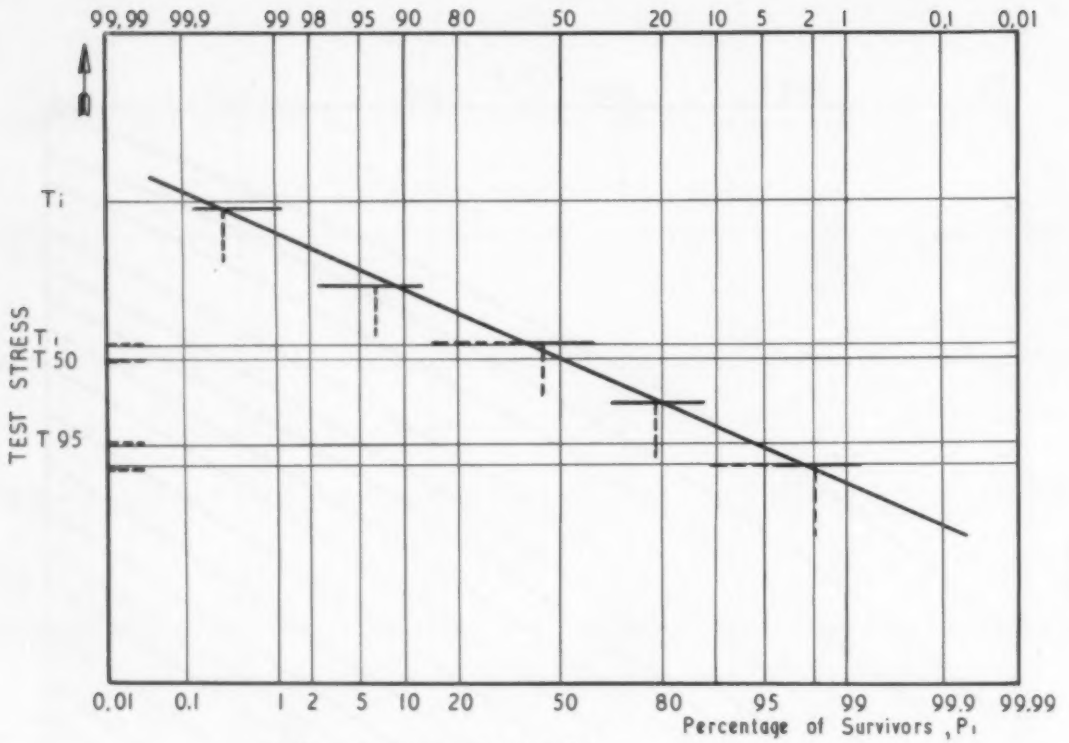
(a)

3—*Curves showing confidence areas for p in a binomial distribution starting with a sample having c/n survivors: (a) Confidence coefficient 1–2α=0·95; (b) confidence coefficient 0·99*



(b)

The figures on the curves are the number n of specimens in the sample. If for a given value of c/n ordinates p_1 and p_2 are obtained, there are 95 chances in 100 that the true unknown value of p is between p_1 and p_2



4—*Tests at stresses near the fatigue limit*

The short straight lines indicate the areas of 90% confidence at the different percentages

cycles, on the assumption that a piece which has not broken by then will never break. This number of cycles must be regarded as characteristic of the tests, and should be stated at the head of the results. The statistical estimates envisaged are based on the percentage of survivors in each sample, and have a binomial distribution.

Let p be the proportion of survivors, and $q=1-p$ the proportion of broken specimens.

Consider a first sample with n_1 test pieces, stressed from 0 to a tension T_1 giving the result $p_1=0{\cdot}45$.

The confidence interval for p at 95% or 99% is found from binomial distribution graphs such as shown in Figs. 3*a* and 3*b*.

Consider also other samples giving p_i for the tension T_i using n_i specimens. (It is advisable for the numbers of test pieces to be not widely different.)

The results are plotted on probability paper, the applied stresses T_i being plotted as ordinates, with the proportion of survivors p_i being plotted as abscissae, with their confidence intervals (Fig. 4).

Exact results can be obtained from this diagram by statistical methods which are too involved to explain in this paper. It will often be accurate enough to draw a straight line through the points. From it, the value T_{50} is then deduced, for which 50% of unbroken specimens can be expected. Similarly T_{90}, T_{95}, T_{10} can be obtained, but the confidence intervals of these estimates can be calculated only by statistical methods beyond the scope of this article.

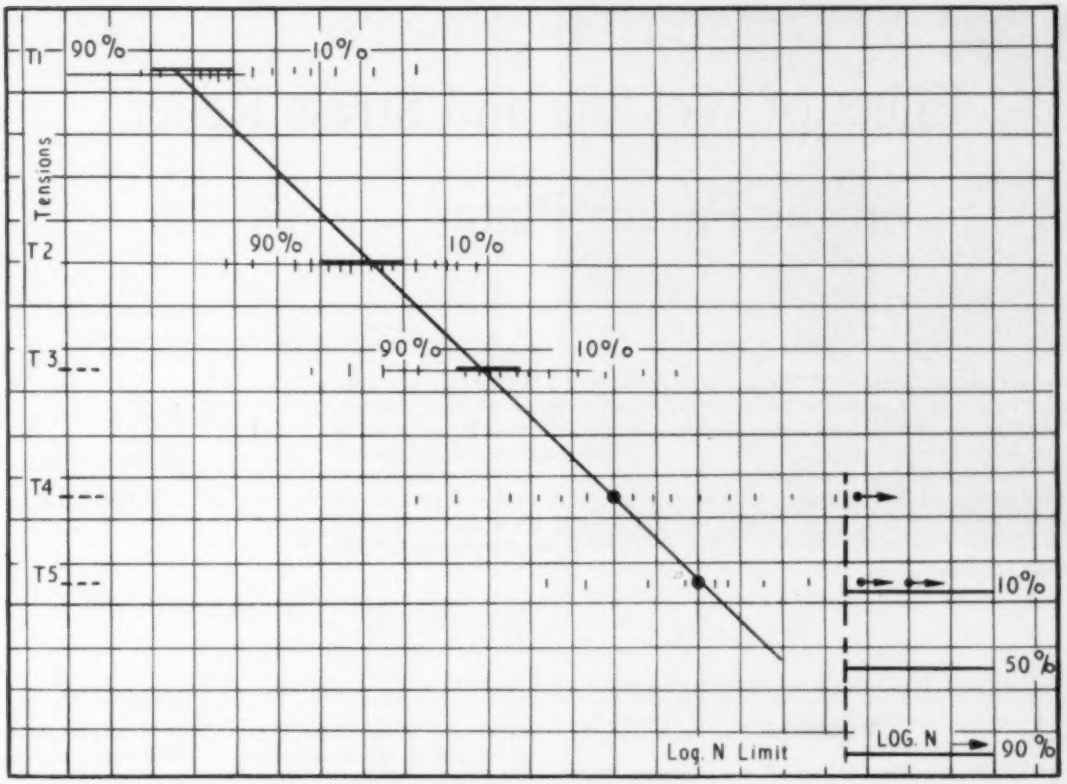
The sharp relationship between N and T represented by a line on the Wöhler diagram is blurred by a third variable, the probability of failure. Knowledge of this probability is absolutely necessary to the engineer who wishes to apply in practice what he finds in the literature of the subject.

Zone (iii) Tests at Stresses slightly Above the Endurance Limit

It has been shown how to estimate the number of survivors above N_{limit}, and how to draw the diagram to the right of the preceding dotted line. Is it not possible from the same tests to obtain complementary information on the part to the left of the dotted line? In reality, some of the test results are undetermined, and do not enable the mean, the standard deviation, and so on, to be calculated.

The arithmetic mean can be replaced by the median; that is, the value above which 50% of results are found, using the linear-Gaussian diagram like Fig. 2.

In fact, this is all the information that can be obtained, although several authors doubt whether the distribution in zone (iii) is normal, and it would thus



5—*Graphical representation of results on a Wöhler diagram*

The thick lines indicate the areas of 90% confidence in the mean values. The thick dots are medians. The figures show the percentage of survivors. The short vertical lines are the co-ordinates of each test result. Arrows indicate unbroken specimens

be unsafe to use criteria which are only true for a normal distribution.

Graphical Representation of the Results on a Wöhler Diagram

It is advisable, at each chosen level, that the maximum information should be stated on the diagram. Each point on the diagram (Fig. 5) represents one test piece. The unbroken test pieces are indicated by an arrow pointing in the direction of increasing N. At each stress, the confidence zone for the mean is shown (90%, 95%, or 99% at will, but the figure must be stated), also the confidence interval for the whole population must be plotted. Near the fatigue limit, the medians are plotted to the left of N_{limit} as well as the percentage of survivors. This diagram should be supplemented by one like Fig. 4.

What has been stated is still true even if the lower limit of stress is not zero, but this fact should be stated on the diagram.

Effect of Welding and Stress Relief on the Parent Plate

The mechanical properties of the heat-affected zone of a welded plate are compared with those of the parent plate; for mild steel, notch-ductile steel, and low-alloy high-strength weldable steels, the properties of the H.A.Z. are equal to those of the parent plate, provided that there is no cracking, or excessive hardening which may lead to cracking.

Initiation of brittle fracture in a welded structure is shown to be strongly dependent on the presence of residual stresses; the use of notch-ductile steels considerably reduces the risk of such fractures even in the presence of residual stress.

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M.ENG., A.I.M.,
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Introduction

THE theme of the 1958 I.I.W. open session, 'Welding in the Chemical Industry', included the influence of welding, preheating, and the like, on the properties of welded components in comparison with the properties of the parent material; this paper deals with the latter aspect, i.e., the effect of welding on the parent material. As steel is probably the most common material used in chemical plant (and indeed in many other forms of equipment), it is intended to deal with the effect on steel plates of welding and stress relief. It is obviously of interest to the chemical engineer to know whether the properties of the parent plate are impaired by such treatment and, particularly, whether the advantages of the various improved types of steel are maintained in a welded fabrication.

In discussing the effects of welding on the parent plate, two major aspects can be considered.

Firstly, there are changes produced in the plate immediately adjacent to the weld, due to the thermal cycle it undergoes as a result of welding. This cycle involves relatively rapid heating and cooling, the maximum temperature reached at any particular point in the plate being dependent on the welding technique employed and on its distance from the weld (*see* Fig. 1). The steel nearer to the weld is heated into or above its transformation temperature range, and the subsequent rapid cooling leads to appreciable hardening in this region, which now has a different metallurgical structure from the body of the plate; the quenched structure is revealed on etching as a darker zone (Fig. 2). It is this region which is usually described as the heat-affected zone, or H.A.Z. In addition to hardening, there is also a liability to crack formation in the heat-affected zone; the incidence of such cracking is related to the temperature at which metallurgical transformation is completed during cooling, and to the amount of hydrogen which has diffused from the weld metal into the parent plate.[1] The ease of production of metallurgically sound joints, *i.e.*, joints free from cracks or excessive hardening of the heat-affected zone, is the generally accepted criterion of weldability in Great Britain. This criterion is associated with the type of steel considered, rather than with the particular application for which the steel is being used.

Secondly, there is the overall behaviour of a welded structure, which determines whether welding can be successfully used for a particular application. This behaviour is affected by the metallurgical changes already described, but also by other factors, such as the influence of residual welding stresses, design of components, service environment, and so on. Because of the many factors governing suitability for service, it is likely that every structure, or at least every type of structure must be treated separately, and no test of the 'weldability' of a steel in this wider sense will be of universal application. The latter part of this paper deals with one part of this problem—the effect of residual stresses and various stress relieving processes on the behaviour of welded plates under applied loads. It will be shown that the notch ductility of the plates, frequently included in Continental definitions of weldability, has a considerable influence on their behaviour.

Paper presented at the open session of the I.I.W. Annual Assembly at Vienna, in June 1958.

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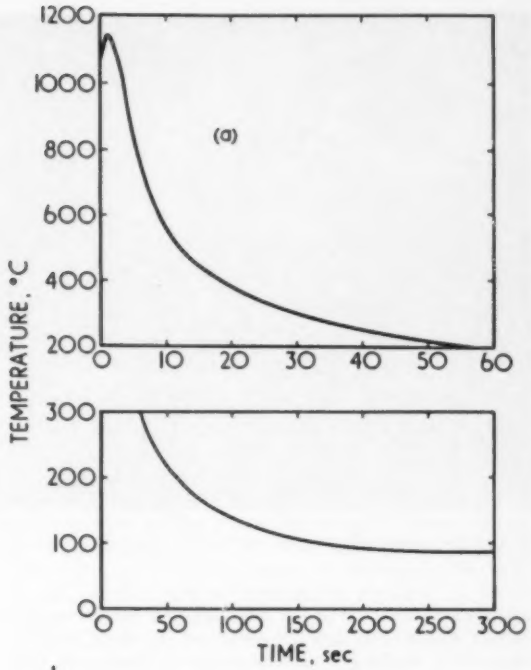
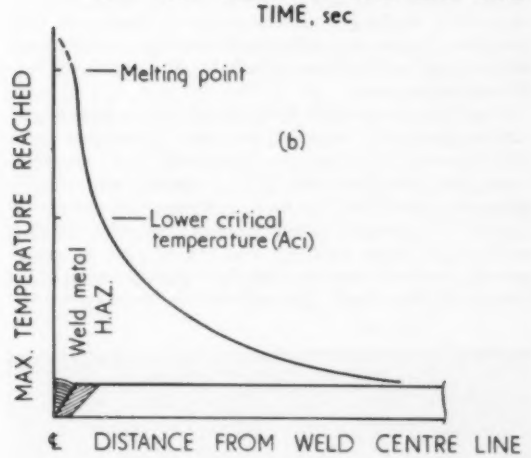
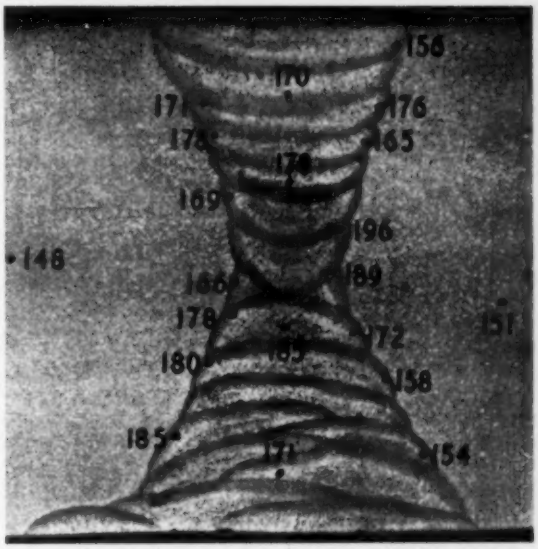




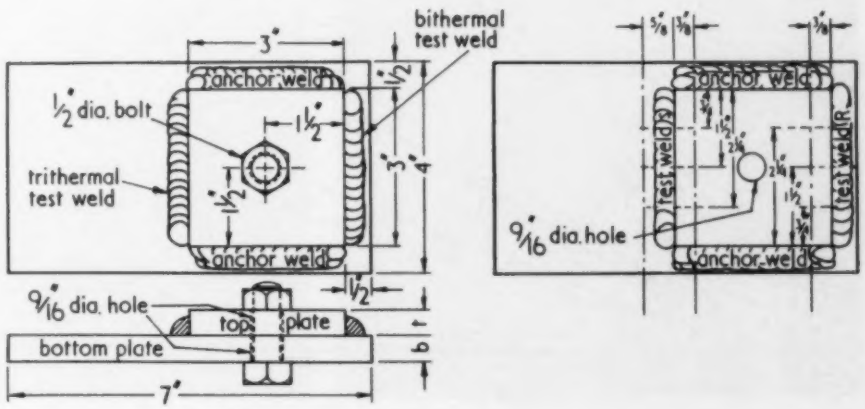
1—*Temperature distribution in welds: (a) typical thermal cycle in heat-affected zone of filler weld (after Cottrell, Jackson, and Whitman)[11]; (b) relationship of maximum temperature and distance from centre of butt weld*



2—*Macrograph of weld in 3 in. thick steel, showing heat-affected zone hardnesses*

Metallurgical Properties of Heat-Affected Zones

The hardening and liability to cracking of heat-affected zones varies in different classes of steel. The interest of the chemical industry in this aspect of the behaviour of steel is considerable; the use of the higher-tensile steels is increasing in importance for chemical processes carried out at increased working pressures and temperatures, for which extreme thicknesses of plates would be needed if mild steel were used for the vessels.



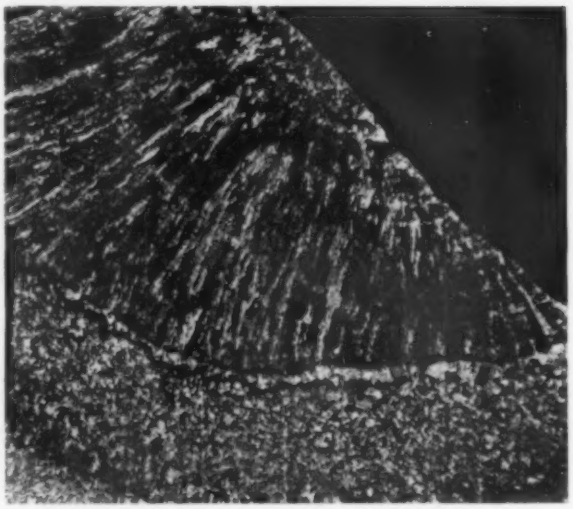
3—*C.T.S. test assembly for low-alloy steels*

Weldability tests

One of the first problems in any assessment of weldability of steels relative to heat-affected zone hardening and cracking is how to test the material. Various tests have been suggested, based on varying the rates of cooling, or the restraint during cooling, but for general application any test must be simple. Two current tests fulfil this requirement: the Battelle underbead cracking test[2] developed in the U.S.A., and the controlled thermal severity (C.T.S.) test[3] developed by the B.W.R.A. The latter test is considered to be the most useful at present available, although it is not perfect and is being examined further by the I.I.W. Restrained fillet welds, with varying cooling rates, are used as the test welds, *see* Fig. 3. The conventional procedure[3] for comparing different steels is to lay a ¼ in. fillet as the test weld, and to examine three sections cut from this. The thermal severity can be altered by varying the plate thickness, and is calculated from the selected heat-flow paths perpendicular to the weld, expressed in multiples of ¼ in. In the example of Fig. 3, weld *R* has a thermal severity of 8 (*i.e.*, 1 in. + 1 in. = 4 × ¼ in. + 4 × ¼ in.), and weld *S* a thermal severity of 12 (*i.e.*, 1 in. + 1 in. + 1 in.). Similar calculations are made for any other combinations of plate thicknesses. The highest thermal severity at which crack-free welds can be produced is taken as an indication of the relative weldability of the steel. In addition, measurements of the maximum hardness likely to be encountered in practice can be obtained from the weld sections. The rates of cooling in the heat-affected zones of test welds are plotted in Fig. 4; these agree well with measurements made on joints in full-scale structures.[1]

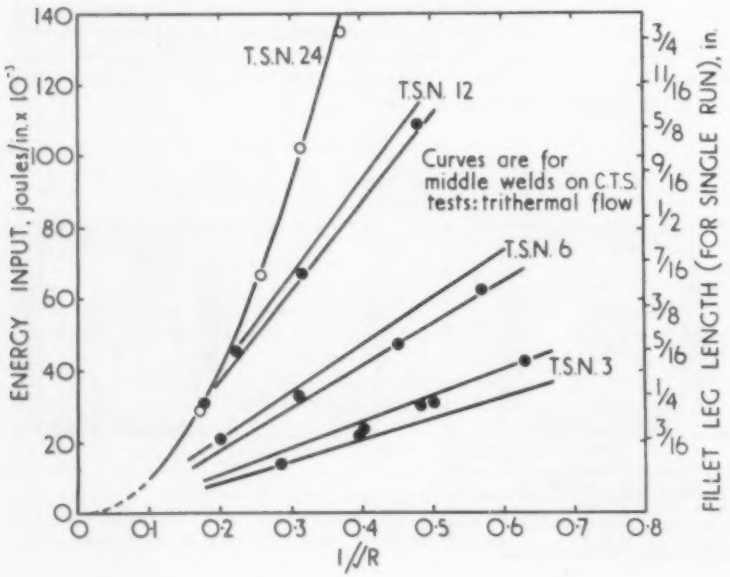
High-strength steels

Mild steel, containing about 0·15–0·20% carbon and 0·4–0·6% manganese, is probably the commonest structural material that exists, and fortunately presents no problem in weldability. The maximum hardness usually encountered in the heat-affected zone does not exceed about 250 D.P.N., and no cracking occurs, even when welding is carried out with electrodes that introduce large quantities of hydrogen. However, when a steel of higher strength is required, problems of weldability arise.



5—Cracks in heat-affected zone of 0·041% C–0·73% Mn steel ×20

If increased strength is obtained by increasing the carbon content of the steel, it is soon found that heat-affected-zone cracking is appearing; for example, 1 in. plate of steel with 0·41% carbon and 0·73% manganese is liable to such cracking, as shown in Fig. 5. It is possible to avoid these cracks by suitable preheating before welding, and taking care to maintain the preheat temperature throughout the welding operation, but such procedures are a nuisance to



4—Rates of cooling of heat-affected zones of welds in C.T.S. test (T.S.N. = thermal severity number, R = rate of cooling at 300° C., °C./sec)

Table I

Composition and properties of a low-alloy high-tensile weldable steel ('Ducol W.30')

Composition, % (max.)		*Composition, % (max.)*	
C	0·17	Cr	0·70
Si	0·30	Mo	0·28
S	0·05	V	0·10
P	0·05	Ni	0·30 residual
Mn	1·50	Cu	0·30 residual

Thickness, in.	*0·5% P.S. tons/sq.in. (min.)*	*U.T.S., tons/sq.in.*	*Elongation, % 8 in. Gauge Length*	*Elongation, % 2 in. Gauge Length*
Under $\frac{3}{4}$	30	38–45	10	—
$\frac{3}{4}$–1	30	38–44	15	—
1–3	27	36–42	15	60 − U.T.S.
3–6	24	34–40	13	

Table II

Properties of plates and welds (3 in. thick 'Ducol W.30')

Specimen	*Yield Stress, tons/sq.in.*	*U.T.S., tons/sq.in.*	*Red. in in Area, %*
As-welded:			
Parent plate	27·8	36·9	61
Transverse weld tensile	30·0	36·4*	56
Stress Relieved 2 hr, 650°C.:			
Parent plate	29·0	37·0	63
Transverse weld tensile	31·0	36·5*	63

* Failure in parent plate

apply on a large scale to thin plates. There is also a limit to the extent to which plain C–Mn steels can be used for plates of higher tensile strength while avoiding carbon segregation, and hence areas of very poor weldability. Alternative ways of obtaining higher-strength steels, without sacrificing weldability, have therefore been sought.

The low-alloy high-tensile structural steels have been developed to provide material which is readily weldable and allows considerable economy of steel by suitable design. To maintain weldability, the general practice has been to reduce the carbon content of these steels, as compared with mild steel, and to add small amounts of alloying elements to produce the higher strength required.

The composition and properties of one of the more recent of such steels, 'Ducol W.30', are given in Table I; this was developed to give its higher tensile strength in plates up to 6 in. thick. The metallurgical development of this steel, and its application to high-duty welded pressure vessels, has been described in some detail.[4,5] It was found that the expectation of good weldability was fully justified, provided that moderate precautions were taken in the way of pre-heat, or in the use of low-hydrogen electrodes (for thinner plates). Figure 6 shows macrographs of a weld in 3 in. thick plate, as welded and after stress relief, together with the results of a hardness survey of weld and heat-affected zone. Results of transverse tensile tests on such welds are given in Table II; these show that there is no difficulty in producing welds of adequate strength to match the plate, and also that the properties of the plate, and of the composite samples, are unaffected by stress-relieving treatments.

Further evidence of the weldability of this steel is given by the results of the C.T.S. tests, for a series of 'Ducol W.30' plates covering the range of compositions likely to be encountered in normal production, given in Table III. All these tests were conducted

6—Weld in 3 in. thick low-alloy steel ('Ducol' W.30), showing hardness values: (a) as welded; (b) after stress relief for 2 hr at 650°C

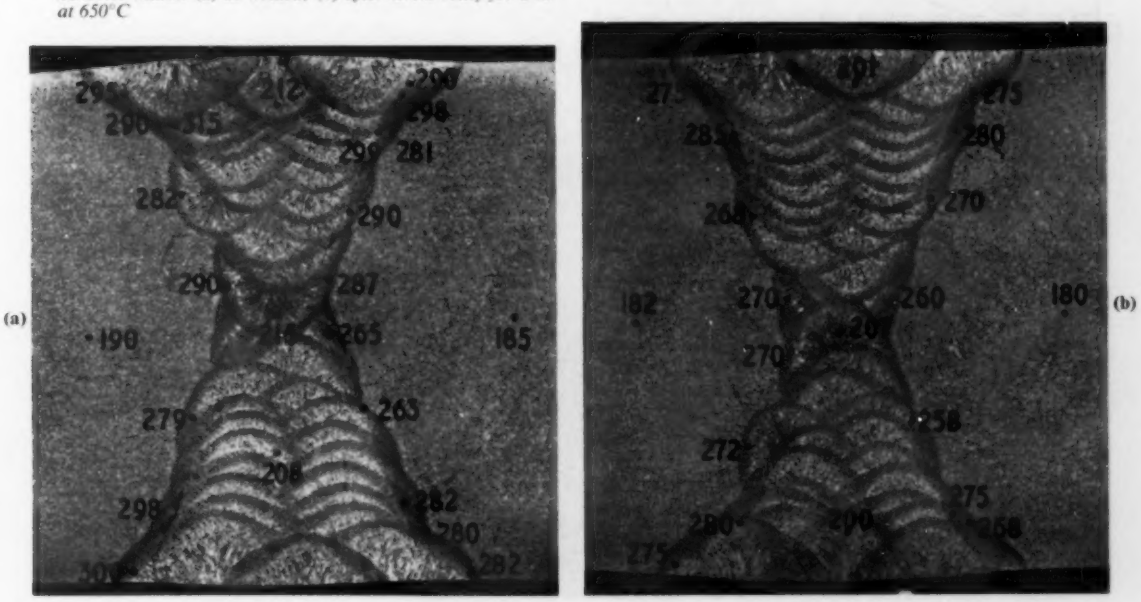


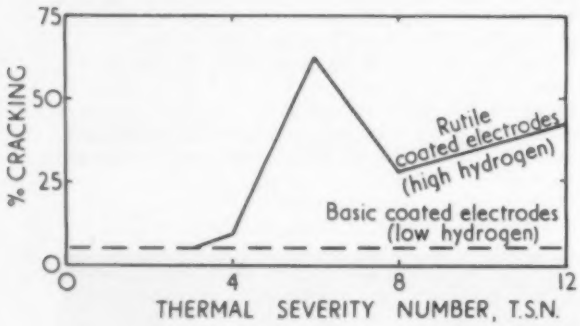
Table III

C.T.S. tests on several casts of 'Ducol W. 30' welded with low-hydrogen electrodes

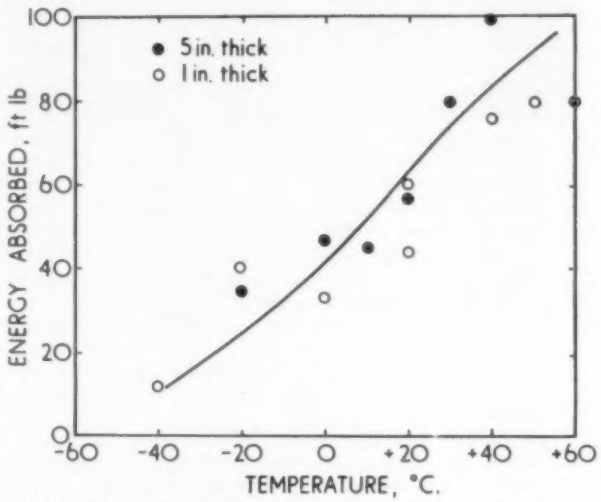
Parent Plate Composition, %										*C.T.S. Test Results*	
C	*Si*	*S*	*P*	*Mn*	*Ni*	*Cr*	*Mo*	*V*	*Cu*	*T.S.N.8*	*T.S.N.12*
0·11	0·17	0·022	0·050	1·46	0·11	0·42	0·24	0·10	0·12	No cracks	No cracks
0·135	0·15	0·036	0·022	1·32	0·10	0·69	0·26	0·08	0·11	,,	,,
0·145	0·15	0·023	0·014	1·20	0·10	0·58	0·26	0·07	0·13	,,	,,
0·15	0·13	0·032	0·015	1·44	0·08	0·62	0·29	0·08	0·12	,,	Very small cracks on one section (2%)
0·155	0·18	0·038	0·015	1·24	0·09	0·49	0·28	0·07	0·15	,,	No cracks
0·155	0·15	0·030	0·043	1·58*	0·08	0·52	0·28	0·07	0·13	,,	Small cracks (average 7%)

* Manganese is above specification maximum in this plate

using a low-hydrogen electrode, with no preheat, and all results are good; but with the rutile type of electrode coating, cracking generally occurs above a thermal severity of T.S.N. 4, as shown in Fig. 7. The indications of the tests are therefore that (*a*) the steel has a fairly constant weldability, and (*b*) the use of low-hydrogen electrodes is essential if plates up to 1 in. thick are to be welded without preheat. For greater thicknesses, some degree of preheating is essential, and for heavy plates (5–6 in.) preheating may be up to 200°C. Under shop fabricating conditions, this does not usually present any great difficulty, and certainly many vessels have been successfully made from



7—*Results of C.T.S. tests on 'Ducol' W.30 steel, using rutile and basic-covered electrodes*



8—*Charpy* **V***-notch energy curve for low-alloy high-tensile weldable steel ('Ducol' W.30)*

9—*Three 35 ft dia. butane storage tanks fabricated in 'Coltuf' 28 notch-ductile steel*

'Ducol W.30' in this way, using both automatic and manual welding.

The notch ductility of the steel is good, as is shown by the Charpy **V**-notch curves in Fig. 8.

Notch-ductile steels

Mild steel is, under certain circumstances, liable to fail suddenly in a brittle manner in large structures, and improved qualities of steel have been developed to lessen the risk of such failures. In Britain, a series of notch-ductile steels has been developed, of successively increasing notch ductility, specified in BS.2762. These are being used in the chemical industry for various types of storage vessels and plant for use at low temperatures. A typical example is the group of 35 ft dia. butane storage tanks shown in Fig. 9, made from 'Coltuf 28' (*ND*. IV) steel. All the '*ND*' steels are characterized by having a low carbon content, with an increased manganese content to maintain the tensile strength in the range of 26–33 tons/sq.in. From the metallurgical aspect, weldability is therefore exceptionally good. However, some doubts had been expressed about the effect of the high manganese contents (up to 1·4%) on the liability to heat-affected-zone cracking; a series of tests was therefore conducted, using both the Battelle underbead cracking, and the C.T.S. methods, on a series of typical notch-ductile steels, together with a few steels of exceptionally high carbon and manganese contents. The results have already been published;[8] the conclusions reached were that, first, the C.T.S. test was more severe than the Battelle underbead cracking test, and further, that within the range of allowable compositions, weldability of the *ND* steels was confirmed as excellent.

Mechanical Properties of Composite Structures

In mild steel, welding produces no significant metallurgical change, as shown by the data in Table IV for tensile strength and notch ductility. However, even in mild steel a weld, by reason of the residual stress associated with it, can act as the origin of a brittle crack, occasionally with disastrous consequences, such as the complete collapse of the oil storage tank shown in Fig. 10; means of reducing the danger of such cracks have therefore been sought. Similarly, the presence of a weld, even when all reinforcement is removed and the surfaces are ground, can reduce the fatigue strength of mild steel shafts[7] and similar components. Indiscriminate and uneducated use of welding can therefore lead to disappointing results in service.

Effect of stress-relieving treatments

Table IV shows that stress relief has little effect on the impact properties of the plate and heat-affected zone. It has been asked whether such treatment is really of any benefit, at least in reducing the danger of brittle fractures. It is well known that, in the as-welded state, high residual tensile stresses exist along the axis of the weld (up to the yield stress) and for some distance on either side, the magnitude of the

Table IV

Properties of welds in mild and notch-ductile steels

Impact properties

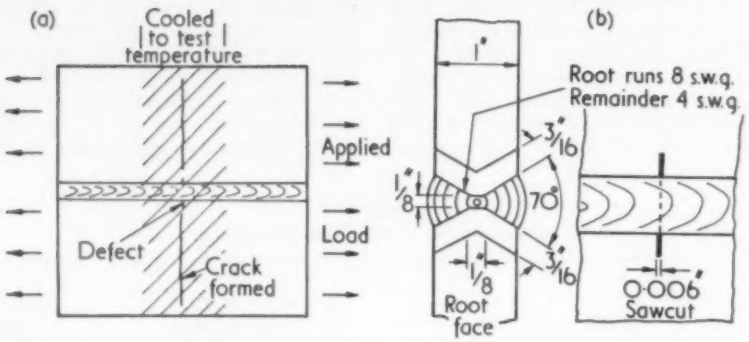
Quality	*Composition, %*						*Welding Electrode (BS.1719 Class)*	*Parent Plate Impact Properties*				*Heat-Affected Zone Impact Properties*			
								As-welded		*Stress Relieved*		*As-welded*		*Stress Relieved*	
	C	*Si*	*S*	*P*	*Mn*	*Ni*		*Temp., °C., for 15 ft–lb.*	*50% Fibre*	*Temp., °C., for 15 ft–lb.*	*50% Fibre*	*Temp., °C., for 15 ft–lb.*	*50% Fibre*	*Temp., °C., for 15 ft–lb.*	*50% Fibre*
Mild steel	0·17	0·03	0·037	0·022	·55	—	E.219	+20	+55	+ 5	+45	−15	+10	−20	+10
ND. I	0·115	0·04	0·037	0·023	1·01	—	E.219	−20	+15	−10	− 5	−55	−35	−65	−30
ND. II	0·115	0·04	0·037	0·023	1·01	—	E.317	−30	+10	−15	+10	−65	− 5	−50	−25
ND. IV ('Coltuf 28')	0·145	0·15	0·025	0·026	1·12	—	E.616, dried at 150°C.	−65	−25	−55	−20	−80	−60	−75	−60
'Coltuf 32'	0·18	0·37	0·037	0·017	1·41	0·20	E.614, dried at 150°C.	−55	−25	−65	−15	−80	−45	−65	−50

Tensile properties

Quality	*Parent Plate*								*Transverse Weld*							
	As-welded				*Stress Relieved*				*As-welded*				*Stress Relieved*			
	Yield Stress, tons/sq. in.	*U.T.S., tons/sq. in.*	*Elong. on 2 in., %*	*Red. in Area %*	*Yield Stress, tons/sq. in.*	*U.T.S., tons/sq. in.*	*Elong. on 2 in., %*	*Red. in Area %*	*Yield Stress, tons/sq. in.*	*U.T.S., tons/sq. in.*	*Elong. on 1 in., %*	*Red. in Area %*	*Yield Stress, tons/sq. in.*	*U.T.S., tons/sq. in.*	*Elong. on 1 in., %*	*Red. in Area %*
Mild steel	18·9	28·8	34	62	15·4	27·2	39	60	19·0	29·0	36	61				
ND. I	15·6	26·9	40	70	15·3	25·5	39	60	18·5	27·0	42	68	16·5	26·0	43	64
ND. II	17·0	26·9	40	69	16·1	26·0	40	65	20·0	27·9	41	68	17·0	26·4	45	69
ND. IV ('Coltuf 28')	20·9	31·2	41	69	19·1	29·6	40	68	21·1	31·5	45	74	21·7	30·3	46	74
'Coltuf 32'	22·9	35·0	32	59	22·8	34·0	33	57	24·0	36·2	20*	64	22·4	33·4	26*	67

* 2 in. gauge length used for weld specimens in 'Coltuf 32'

10—*Oil storage tank collapsed after brittle fracture*



11—*Wide-plate crack initiation tests: (a) specimen during tests; (b) weld preparation and saw-cut notch*

stress gradually decreasing, until at about 4 to 6 in. from the weld the residual stress is zero.

The part that such stresses can play in initiating brittle fracture has been demonstrated in a convincing manner in a series of wide-plate tests, described by Kennedy,[8] on welded mild-steel plates 36 in. square. In this test, which has been described elsewhere in detail,[9,10] two plates, 36 in. × 18 in., are butt welded to form a 36 in. × 36 in. square, and a defect is deliberately introduced in the form of a saw-cut, 0·006 in. wide, in the heat-affected zones on either side of the weld at the middle of its length (*see* Fig. 11). The plate is cooled to the test temperature, and then load is applied parallel to the axis of the weld. The test itself is a measure of two things: (*a*) in the presence of a defect in the weld, how easily a crack can be initiated, and (*b*) once the crack has started, whether it will propagate completely across the specimen.

Kennedy[8] used the test to investigate the effect of various treatments applied before, during, or after welding, and the results are reproduced in Table V. This describes as a short crack one which did not extend more than about 8 in. on each side of the weld, whilst full-width fractures are those which exceeded this amount, and usually involved total separation of the two halves of the specimen.

The tests carried out on specimens given a furnace stress-relieving treatment at 650°C. for 1 hr showed conclusively that this treatment provides a high degree of safety against brittle fracture, since a stress of yield-point magnitude had to be applied before such failure took place. It has been argued on occasions that the benefit of the 650°C. stress-relieving treatment is at least partially due to metallurgical effects in softening of the heat-affected zones, as well as to the removal of residual welding stresses. However, in these tests, removal of the stress field round the weld was also achieved by the alternative method of mechanically straining the whole specimen at a temperature above the transition temperature. Such treatment would have no effect on the metallurgical characteristics of the weld and heat-affected zones comparable with that of the 650°C. treatment; indeed, mechanical straining would result in some degree of further hardening of these zones. The fact that for the mechanically strained specimens yield-point loading was again necessary for the initiation of brittle fractures at temperatures as low as —50°C., confirms that the factor which assists this initiation in the as-welded specimens is the residual stress field, and not the metallurgical characteristics of the weld and adjacent plate material.

In the case of stress relieving by the Linde low-temperature process, some degree of additional resistance to the initiation of brittle fracture was conferred, but not to the same extent as by either of the two above-mentioned procedures. It could therefore be suggested that, although it is not fully effective, such treatment may well be worthwhile for structures liable to brittle fracture which are too large to allow furnace stress relief, for instance, ships or storage tanks.

Further confirmation of these suggestions is the fact that successful treatments are those known to afford relief of residual stresses, whereas the unsuccessful treatments are those unlikely to have significantly changed these stresses. The general beneficial effect of furnace stress relieving has been recognized for many years in various chemical engineering codes of practice, which require such treatments for vessels once the plate thickness exceeds a certain minimum value. Service experience with stress-relieved vessels of mild steel, working at low temperatures in chemical plant, confirms the wisdom of this practice and shows that it does afford a considerable safeguard against failure by brittle fracture.

Table V

Results of wide-plate tests on treated specimens*

Treatment	*Specimen*	*Test Temperature, °C*	*Stress tons/sq.in.*	*Fracture*
Furnace treatment ment for 1 hr at 650°C	*b*1	−8	15·0	Not broken
	*b*2	−8	18·4	Not broken
	*b*3	−8	15·7	Not broken
	*b*4	−25	15·0	Full-width fracture
	*b*5	−56	16·6	Full-width fracture
Linde low-temperature process	*e*7	−25	13·0	Full-width fracture
	*e*8	−25	11·0	Full-width fracture
	*e*9	−25	10·0	Full-width fracture
Mechanical yielding				
(at + 30°C)	*d*1	−8	16·1	Full-width fracture
(at + 30°C)	*d*2	−8	15·5	Full-width fracture
(at + 47°C)	*d*4	−22	15·6	Full-width fracture
(at + 40°C)	*d*5	−50	16·8	Full-width fracture
Preheating to 200°C before welding	*g*1	−8	15·4	Not broken
	*g*2	−8	16·4	Not broken
	*g*3	−8	14·6	Full-width fracture
	*g*4	−25	2·7	Full-width fracture
	*g*5	−25	6·7	Full-width fracture
	*g*6	−25	0	Short spontaneous crack
Post-heated 4 hr at 250°C	*c*1	−8	7·8	Full-width fracture
	*c*2	−8	6·5	Full-width fracture
	*c*3	−8	5·0	Full-width fracture
Local post-heating of weld to about 650°C	*f*2	−8	1·3	Short crack
	*f*3	−8	0	Short spontaneous crack

Usefulness of notch-ductile steels

There are occasions on which stress relieving is impossible, and for such cases the notch-tough steels have been developed. As already stated, a range of such steels is available in Britain to standard specifications (BS.2762, *ND* I–IV). Details of the properties of these steels are given in Table VI, which also includes particulars of a proprietary notch-tough steel, 'Coltuf 32', of somewhat greater tensile strength. The steels were developed on the basis of the Charpy **V**-notch impact test, and it was assumed that improvement of notch toughness as indicated by this test would lead to increased safety from brittle fracture in service. The value of such steels would obviously be reduced if it were found that welding impaired their properties (though such an effect was not anticipated), and therefore a series of tests on 1 in. thick plates has been carried out to determine the impact and tensile properties of such plates in the heat-affected zones. The results of these tests are included in Table IV, and

Table VI

Properties of notch-ductile steels (¾–2 in. thick)

Quality		*Tensile Strength tons/sq.in. Min.*	*Tensile Strength tons/sq.in. Max.*	*Yield Stress, tons/sq.in. Min.*	*Min. Elongation,* %	*Charpy Impact Values Test Temperature, °C.*	*Charpy Impact Values Min. Average 3 Test Pieces, ft–lb.*	*Charpy Impact Values Min. Individual Test Piece, ft–lb.*
ND. I	*A*	26	31	14·25	20	0	20	15
	B	28	33	15·25	20			
ND. II	*A*	26	31	14·25	20	−15	20	15
	B	28	33	15·25				
ND. III	*A*	26	31	14·25	20	−30	20	15
	B	28	33	15·25				
						−10	—	45
ND. IV	*A*	26	31	14·25	20	−20	—	40
	B	28	33	15·25		−30	—	35
						−40	—	25
						−50	—	20
'Coltuf 32'		32	38	21	23	−10	—	35
						−50	—	15

* Test piece *A*, 8 in. gauge length

show that the parent plate properties are at least maintained, in both as-welded and stress-relieved conditions.

A further check on the benefit of the notch-tough steels is provided by wide-plate tests of the type already described. Under certain conditions, complete fracture can occur at low applied stresses, but it is found that the maximum temperature at which such low-stress failures can occur falls as the notch-toughness of the steel increases. Although all varieties of notch-ductile steels given in Table VI have not been tested, the results obtained so far provide the following comparison:

Material	*Max. Temp. for Low-Stress Failure, °C.* *Partial*	*Complete*	*Temp. for 15 ft–lb. Impact Value in Charpy V-notch Tests °C.*
Mild Steel	+8	?	+18
ND I	−5	−8	− 8
ND IV	−30	below −60	−60

It will be seen that the improved steels show a better behaviour than mild steel, although there is no direct relationship between the Charpy energy transitions and the wide-plate temperatures quoted.

Fabrication practice and safety in service

Safeguards against the initiation of brittle fracture of welded structures can therefore be provided either by the use of a suitably notch-tough steel, or by efficient stress relief, and the ultimate decision on which course to adopt must be governed by practicability. For example, a chemical reaction vessel which is shop-fabricated and given full stress relief is normally perfectly safe when made in mild-steel plate; but a ship destined for service in regions where low service temperatures are expected cannot be fully stress-relieved, and a notch-tough steel is desirable in the critical portions of the structure. It should be noted, however, that residual stresses only affect initiation of fracture, and have no influence on propagation once such a crack is under way. Propagation will be governed by a combination of the factors of plate temperature, stress, and inherent characteristics of the steel; this is outside the scope of this paper, but it may be remarked that the tougher steels obviously give greater safety in this respect.

For exceptional circumstances, it may be desirable both to use a tough steel, also to stress relieve the structure; examples are provided by large pressure vessels for nuclear reactors, where, although stress relief is eventually to be applied, construction will continue for many months, with a wide range of climatic conditions. The use of a notch-tough steel guards against the danger of fracture during fabrication, and stress relief on the completion of fabrication gives the maximum possible safety to the structure when it goes into service.

A further advantage is obtained from the use of notch-ductile steels for site erection, where welding procedures cannot be closely controlled. Under these circumstances, the good weldability of the steels (due largely to their relatively low carbon content) is beneficial, and their ability to suffer appreciable deformation without becoming brittle, and so initiating cracks from minor weld defects, ensures that no troubles arise from these sources. Such steels have been widely used for penstocks for hydro-electric schemes, and in fact the notch-ductile steel of slightly higher tensile strength ('Coltuf 32') was developed from the *ND* IV steel for such applications.

Table VII

Properties of 1½-in. thick, low-alloy, quenched and tempered steel plate and welds

Plate Composition, %

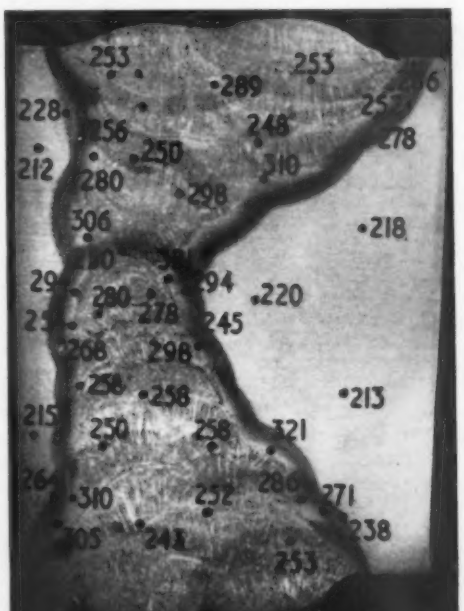
C	Si	S	P	Mn	Ni	Cr	Mo	V
0·14	0·11	0·027	0·013	1·18	0·32	0·78	0·36	0·085

Specimen	*Yield Stress tons/sq.in.*	*U.T.S., tons/sq.in.*	*Elong. on 4√A, %*	*Red. in Area, %*	*Impact Charpy Value at −40°C., ft–lb.*	*Impact Properties Impact Temp. for 15 ft–lb., °C*
Plate (water-quenched and tempered)	38	44	26	68	34	−65
Heat-affected zone	—	—	—	—	75	−70
Transverse weld tensile*	38	44	21	66	—	—

* Failure in the plate material

C.T.S. Test (Class 6 electrode)

T.S.N.	*Cracking*	*Hardness of H.A.Z., D.P.N.* *Max.*	*Average*
10	None	391	337
12	None	401	391



12—*Butt weld in 1½ in. plate of quenched and tempered steel, made manually with low-hydrogen electrodes (preheat 100°C.), showing hardness values* ×2

'Super' steels

The aim of subsequent developments has been to produce steels which possess even higher yield and ultimate strengths than those already described while having the same, or even better, weldability, and in addition having improved notch toughness. The advantages of such a steel are obvious; its use would make it possible to reduce the weight and thickness of plates, the time taken to weld a structure, and, particularly for large vessels which must be site welded, the danger of cracking and brittle fracture, either during fabrication or in service.

Such steels are generally in the development stage, but it seems clear that they will derive their properties from a combination of alloying additions and heat-treatment, the latter taking the form of quenching and tempering. They will be supplied in the fully heat-treated condition, and will pose a number of problems for the fabricator as far as forming the plates is concerned, as this must be done at temperatures below the tempering temperature. Not much information is available yet in regard to weldability and the properties of the welded joint, but it would appear that the major problem is not the effect of welding on the parent plate, but the properties of the weld itself.

Evidence of the generally good behaviour of the quenched and tempered steel is given by the results of C.T.S. tests, which are given in Table VII together with the properties of the parent plate. The impact properties of plate and heat-affected zone are also included, and Fig. 12 shows a macrograph of a butt weld in 1½ in. thick plate, with hardness survey results. In all these tests, the welds were made without preheating. It is thus seen that weldability is as good as that of the normalized high-tensile steel, and that the properties of the plate are unimpaired by welding.

Conclusions

To sum up, the points which are of interest to the designer and operator of chemical plant are as follows:

(1) Welding rarely causes a reduction in notch ductility or tensile strength in the heat-affected zone of normal structural steels, either of the specially notch-tough, or the low-alloy high-tensile weldable varieties.

(2) Stress relieving has an extremely beneficial effect in reducing the liability to initiation of brittle fracture in a structure.

(3) The notch-ductile steels permit a greater latitude in the control of welding conditions for site work than does a relatively brittle steel of the same tensile strength.

(4) High-strength notch-tough steels, produced by quenching and tempering, and the addition of suitable alloying elements, also possess good weldability, and are worthy of further development.

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TRANSLATIONS FROM THE RUSSIAN

The British Welding Research Association is undertaking the latest cover-to-cover translation from the Russian. Regular translation of *Avtomaticheskaya Svarka* (Automatic Welding) will start with the January 1959 number, which is expected to arrive in Britain in March, and to be distributed in translation to subscribers in May or late April.

Avtomaticheskaya Svarka is the only monthly journal in any language with the title "Automatic Welding", though the title is not absolutely descriptive except in so far as manual techniques are not emphasized. Equipment and methods of welding are dealt with, as well as metallurgy, the strength of welded work generally, and fatigue testing.

The contents of the August 1958 number, which is being translated as a specimen, are: an editorial summary of the U.S.S.R. 7-year plan (1958–1965) as it applies to welding, with 15 signed articles, three of them on electro-slag welding:

Diffusion Welding in Vacuum, *by Kazakov.*
Current Sources for Electro-slag Welding, *by Gupalo.*
Influence of Type and Polarity of Current On Desulphurization of the Molten Metal in the Electro-slag Process, *by Medovar and Latash.*
Adding Electro-slag Metal to Form Ingots or Shaped Castings, *by Belous and Dudko*
Metallographic Investigation of Welds in Aluminium, *by Grabin and Rabkin.*
Simultaneous Spectrographic Determination of Carbon and other Elements in Steel and Iron, *by Demyanchuk.*
Impact Strengths of Doors for Standard Open Rail Wagons of 60 and 93 Tons, *by Kazimirov, Olifer, Ivanushkin, Kapustyanov, Svinarenko, and Tyagun.*
Primary Weld Metal Crystallization in Fusion Welding, *by Razikov.*
Some Fatigue Strengths of St 3 Rimming Steel under Pulsating Load, *by Bat.*
Effects of Silicon and Manganese on Weld Formation in Single Side Welding, *by Pashchenko.*
Effect on Residual Stresses of Local Heat Treatment of Transverse Butt Welds in Tubes, *by Shevernitskii and Kovtunenko.*
Static Characteristics of the Arc in Gas-shielded Welding with Consumable Electrodes, *by Verchenko.*
Electrical Circuits of Resistance Welders for Thin Sheet, *by Zaitsev.*
Automatic Welding of Aluminium Containers of 2 cu.m Capacity, *by Dovbishenko and Sventsinskii.*
Use of 9-kW Surplus Aircraft Generators for Welding, *by Timofeev.*

The D.S.I.R. translation programme is planned in conjunction with a similar programme in the U.S.A. so that the two shall not overlap, and the maximum of information will thus become available in English. The specimen translation will shortly be available on loan to anyone who writes to the Publications Officer at 29 Park Crescent, London, W.1. The subscription price will be £10 10*s.* 0*d.* for the 12 issues of 1959.

By C. G. Mukhin and N. Y. Palchuk

Niobium in Stainless Steel Weld Metal

The effect of up to 3% niobium on the structure and mechanical properties of welds in 18/8 type austenitic steel was investigated, both in the as-welded state, and after water-quenching, 'stabilizing-annealing', and heating at 650°C. for 1000 hr. It is shown that after heat-treatment a previously unknown intermetallic phase, designated the N-phase, may appear in welds containing over 2% niobium, in addition to the carbides, the niobide Fe_3Nb_2, and the silicide $Fe_4Nb_5Si_3$, which form during welding. The new phase is similar in structure and properties to the chi-phase formed in 18/8 type steels containing molybdenum.

Introduction

Welds in 18/8 type austenitic steels are alloyed with niobium to prevent intercrystalline corrosion; they normally contain up to 1% niobium. That part of the niobium not combined as carbides is in solid solution. The niobium in solid solution promotes the formation of ferrite, although its ferrite-forming effect is considerably less than that of titanium.[1, 2] Other phases have been found in austenitic Cr–Ni steels containing niobium, in addition to ferrite and the carbides of niobium; these are iron niobide, Fe_3Nb_2,[3] a silicide, $Fe_4Nb_5Si_3$,[4] the complex carbide of niobium and chromium Nb_3Cr_3C, with the lattice constant $a_o = 11{\cdot}49$Å,[5] and also the carbide of tantalum and chromium, Ta_3Cr_3C.

In a niobium-bearing 18/8 steel the ferrite forms part of a characteristic eutectic structure. A eutectic of ferrite and a second phase of unknown structure has been observed in an ingot of 18/8 steel, in the weld, and also in the heat-affected zone after welding and gas cutting.[6] The second-phase segregate retained its eutectic structure in rolled steel[6] and in the weld metal after the ferrite had been dissolved by hardening.[6, 7] Simpkinson,[6] and Emerson and Morrow[7] consider that this second phase is the carbide of niobium, though this was not verified experimentally. The eutectic second-phase segregates may easily originate in the parent metal close to the weld in the heat-affected zone.

The hardening of 18/8 type steel containing niobium from 1260–1370°C. caused the eutectic to form at the austenite grain boundaries.[6] The eutectic segregates were also observed in the heat-affected zone, but only a small proportion of this is heated to such high temperatures.

The authors consider that the eutectic may consist not only of ferrite and the carbide of niobium, but also of other phases, for example Fe_3Nb_2 and ferrite. Medovar[2] considers that a eutectic of ferrite and the niobide of nickel may be formed. Figures have been given for the inter-planar distances of various intermetallic phases whose crystal lattices have not yet been determined.[3, 4]

When niobium-bearing 18/8 type steels are subjected to prolonged heating, the free niobium (*i.e.*, that not linked with carbon in carbides) promotes the formation of sigma-phase, although in this case also niobium is less active than titanium, silicon, chromium or vanadium.[1]

In a study of the appearance of sigma-phase in 18/8 type steels containing niobium, Binder[3] showed that where the niobium content exceeds 1%, another phase, which he considered to be the carbide M_6C, was formed as well as carbides of niobium and sigma-phase. (In the present paper, this has been designated the N-phase).* Where the Nb content exceeded 2%, the niobide Fe_3Nb_2 was formed in the intermetallic precipitates, in addition to the other phases enumerated. The occurrence of sigma-phase and the other components reduces ductility and impact strength; the reduction in impact strength was found to be particularly high in steels containing over 2% Nb.

The possibility that various phases may form in welds stabilized by niobium, both in the welding process and during later operations, is of great practical interest. It has been established[2] that the niobium in welds acts primarily as a carbide-forming agent, and only the excess causes the amount of ferrite in the weld to increase. When the niobium content is increased to 1–1·5%, the danger of hot cracking is reduced even if no additional ferrite-forming elements are introduced into the weld.[2, 8]

The suggestion that the formation of the silicide $Fe_4Nb_5Si_3$ during welding causes hot cracking, and the

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This translation, from *Avtomaticheskaya Svarka* ("*Automatic Welding*"), 1957, No. 2, is the second from this Russian monthly to be published in the *Journal* by the B.W.R.A. Paper translated by Major J. H. Dixon.

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* Since the Russian text gives "H-phase", and "H" in Russian usually means the west European "N", this term has been translated as "N-phase".

recommendation that the silicon content of welds should not exceed 0·3%,[4] have been refuted on both theoretical and practical grounds by welding engineers in the U.S.S.R.[9, 2]

It has been noted that niobium contents of up to 1·16% do not cause the formation of sigma-phase even in multi-pass welding;[3] however, prolonged heating at 600–800°C. reduces the ductility of the weld metal. The reduction in ductility and the great decrease in the impact strength of the weld metal after prolonged heating have been explained by the formation of sigma-phase in welds independently of their niobium content.[2, 3, 10, 11]

In the welding of Cr–Ni austenitic steel of the 18/8 type, with a niobium content of the weld metal of about 2%, the formation in the welds of the niobide Fe_3Nb_2 has been noted;[12] however, the method by which the niobide was identified was not indicated.

Experimental Procedure

The work described in this paper has been carried out to clarify the conditions under which sigma-phase is formed in welds in 18/8 type steel during prolonged heating at a temperature of 650°C. X-ray structural analysis was carried out, using iron and chromium radiation.

The analyses of the welds are given in Table I. Test pieces cut from the welds were subjected to prolonged tempering (up to 1000 hr) at a temperature of 650°C. Before tempering, some of the test pieces were water-quenched after holding at 1100°C. for 2·5 hr. The rest of the test pieces were subjected to stabilizing-annealing (heated for 3 hr at 850°C. and air-cooled) before tempering.

Metallographic sections were electrolytically etched in a solution of 450 g citric acid, 350 g potassium iodide, and 50 ml HCl to 900 ml of solvent, at a terminal voltage of 6 V. Otherwise the research procedure was similar to that previously described by the authors.[13]

Experimental Results

As-welded state

An increase in the weld metal niobium content is accompanied by an increase in the quantity of ferrite; however, the ferrite-forming action of the niobium is relatively weak, and the quantity of ferrite is little higher than that in welds containing 0·7% Nb.

It was not found possible by metallographic means to establish the heterogeneity of the ferrite, or to find any eutectic second-phase segregates within the ferrite regions, even where the weld niobium content was high. However, by X-ray structural analysis of intermetallic precipitates obtained from welds containing 2·16% and 3·05% Nb, the latter were found to contain the new phase, which according to Binder[3] is the niobide Fe_3Nb_2, as well as the carbide of niobium. This phase was not found in the precipitate from a weld containing 2·16% Nb, which contained the silicide $Fe_4Nb_5Si_3$ as well as the carbide of niobium (the presence of the silicide was established by five lines on the X-ray photograph). The difference between the phase compositions of the welds is explained by the different silicon contents; the silicon content was higher in the weld containing 2·16% Nb, leading to the formation of the silicide, whilst no silicide was found in the precipitate in welds containing 3·05% Nb and lower silicon.

Increase in the weld niobium content lowered the ductility and impact strength (*see* Table II). The reduction in ductility was caused not only by increase in the quantity of ferrite but also by the formation of the new and brittle structural components.

Table I

Chemical compositions of welds in 18/8 type steel

Weld	*Analysis, %*					
No.	*C*	*Si*	*Mn*	*Cr*	*Ni*	*Nb*
1	0·09	0·75	0·68	19·20	10·64	0·7
2	0·10	0·37	1·45	16·60	9·10	1·32
3	0·12	0·64	1·06	16·99	9·10	2·16
4	0·10	0·50	1·47	17·65	9·23	3·05

Post-welding treatment (tempering)

Sigma-phase appears in welds containing up to 2% Nb during heating at 650°C. for 25 hr. The sigma-phase particles form in the ferrite regions and retain the orientation of the ferrite. Prolonged heating of welds containing over 2% Nb at 650°C. results in the appearance of a new phase, which metallographic examination shows to be greatly reminiscent of sigma-phase. This phase, which will be designated the N-phase, also forms in the ferrite regions and is harder than austenite.

By means of X-ray structural analysis the following phases were identified: Carbide of niobium, sigma-phase (in welds with up to 2% Nb) and N-phase (*see* Table III). The results are evidence that hardly any sigma-phase is formed in welds with high Nb contents; the X-ray analysis of intermetallic precipitates revealed only traces of sigma-phase. Comparison of the analysis figures with published results confirms that the N-phase is identical with the M_6C carbide located under similar conditions in 18/8 type steel containing niobium.[3]

The niobide Fe_3Nb_2 and the silicide $Fe_4Nb_5Si_3$ were not found. The small diameter of the chamber (57·2 mm) did not permit the resolution of a number of lines situated close together on the X-ray spectrum, but the fact that the X-ray pictures did not reveal the niobide Fe_3Nb_2, or the lines of phases other than those for the carbide of niobium and the N-phase, gives a basis for the assumption that these absent phases are dissolved during tempering in the process of the formation of N-phase.

Stabilization at 850°C. for 3 hr, with air cooling

The stabilizing-annealing of welds with niobium contents up to 2% has little effect on their structure or mechanical properties; some spheroidization of the ferrite particles may be accompanied by alteration in the concentrations of alloying elements in the ferrite. However, notable changes are observed in welds containing over 2% of niobium. After stabilization the

Table II

Mechanical properties of welds containing niobium

1 kg/sq.mm = 0·63 tons/sq.in.

Weld No.	*Nb Content, %*	*Heat Treatment*	*U.T.S., kg/sq.mm*	*Yield Point, kg/sq.mm*	*Elong., %*	*Red. Area, %*	*Impact Strength kg.m/sq.cm*
1	0·7	Nil	66·6	34·5	43·2	42·8	0·136
		Stabilized	70·0	33·1	—	—	0·114
		Water-quenched	66·6	25·8	53·9	59·0	0·166
2	1·32	Nil	56·7	23·9	31·0	33·1	0·099
		Stabilized	54·0	24·5	32·5	22·3	0·102
		Water-quenched	57·4	23·1	39·2	36·3	0·121
3	2·16	Nil	56·8	29·2	19·3	15·1	0·084
		Stabilized	56·0	24·0	—	—	0·068
		Water-quenched	61·1	22·7	47·1	41·9	0·114
4	3·05	Nil	56·0	27·1	8·0	15·2	0·075
		Stabilized	—	—	—	—	0·049
		Water-quenched	62·2	24·8	41·7	28·7	0·126
	After 1000 hr tempering at 650° C.						
1	0·7	Nil	69·4	31·0	44·3	37·6	0·096
		Stabilized	70·2	31·0	40·0	38·3	0·099
		Water-quenched	65·8	27·4	58·0	53·3	0·138
2	1·32	Nil	56·1	24·2	24·0	19·4	0·030
		Stabilized	58·3	22·3	27·7	20·7	0·045
		Water-quenched	52·6	23·0	20·7	29·5	0·074
3	2·16	Nil	63·8	30·0	16·7	18·6	0·016
		Stabilized	58·8	24·4	22·0	20·3	0·031
		Water-quenched	61·3	27·3	23·1	18·6	0·034
4	3·05	Nil	70·0	29·1	17·3	14·0	0·021
		Stabilized	66·7	25·7	13·7	10·7	0·022
		Water-quenched	61·8	26·1	25·1	19·3	0·042

Table III

Phases in the intermetallic precipitates in welds after different heat-treatments

Weld Niobium Content, %	*Heat-Treatment*		
	As-welded	*Stabilized*	*Quenched*
2·16	Nb(C,N)* + $Fe_4Nb_5Si_3$	Nb(C,N) + N-phase	Nb(C,N) + N-phase
3·05	Nb(C,N) + Fe_3Nb_2	Nb(C,N) + N-phase + Fe_3Nb_2	
	After 1000 hr tempering at 650° C.		
0·7	Nb(C,N) + sigma-phase	—	—
1·32	Nb(C,N) + sigma-phase + trace of N-phase	—	—
2·16	Nb(C,N) + N-phase + traces of sigma-phase	Nb(C,N) + N-phase	—
3·05	Nb(C,N) + N-phase + traces of sigma-phase	Nb(C,N) + N-phase	Nb(C.N) + N-phase

* The carbide of niobium, NbC, can dissolve nitrogen without altering the crystal lattice constant.

impact strength of these welds is greatly reduced (*see* Table II). The weld microstructure is altered, and a considerable quantity of the finely precipitated new phase appears in the austenite.

Investigation of the intermetallic precipitates in the welds confirmed the presence of carbides and of N-phase, which originates as a result of stabilization. The niobide Fe_3Nb_2 is also present in welds containing 3·06% of niobium after stabilization. No sigma-phase was found in welds containing over 2% Nb; thus the reduction in ductility after stabilization is caused by the formation of N-phase.

Tempering of stabilized welds

Tempering at 650°C. results in the formation of sigma-phase in the ferrite in welds containing up to 2% Nb. Deep etching reveals the presence in the weld metal of eutectic segregates, which retain the ferrite orientation, as well as sigma-phase particles. No sigma phase is observed to form with a weld niobium content exceeding 2%.

After 1000 hr of tempering at 650°C. mechanical tests showed a further reduction in the ductility of welds rich in niobium. This is caused by the segregation of a large quantity of N-phase. In this case,

X-ray structural analysis of the intermetallic precipitates revealed the presence of carbides and N-phase. The niobide Fe_3Nb_2, which was observed in the precipitate from weld 4 (Table III) immediately after stabilization, was not found after 1000 hr of tempering.

Comparison of the mechanical properties before and after heating at 650°C. shows that prolonged tempering is accompanied by further reduction in the ductility of the weld metal. The reduction in ductility of welds with niobium contents not exceeding 2% was caused by the formation of sigma-phase; in welds containing over 2% of niobium it is caused by the further segregation of N-phase, above the quantity formed in the process of stabilization at 850°C.

Water-quenching

The microstructure of test pieces after water-quenching was austenite, in which carbides of niobium were dispersed. The changes in the microstructure of welds with high niobium contents are of the greatest interest. In these also, heating results in the solution of the ferrite. Eutectic segregates, which remain after heating at 650°C. and are oriented according to the ferrite, are observed after quenching in welds containing 1·32% Nb.

The X-ray photographs from sections of the quenched test pieces confirm that the ferrite is dissolved on heating, and show austenite lines, and also lines for a phase with a crystal structure about which it would be hard to draw a definite conclusion, though a considerable number of the lines could belong to Fe_3Nb_2 or N-phase.

The X-ray photographs from sections of the quenched test pieces 3 and 4 showed that carbides of niobium and N-phase were present in the two welds. The X-ray plate of the precipitate from weld 4 is distinguished by the disappearance of the characteristic lines of Fe_3Nb_2; from this it may be conjectured that this constituent is dissolved at high temperatures.

Water-quenching increased the ductility of welds. It should be noted that the formation of N-phase reduces the ductility; this was found after stabilization or after prolonged tempering at 650°C. However, the overall effect of the two processes—the homogenization of the weld metal and the formation of N-phase as a result of heating—results in a moderate increase in the ductility of the hardened weld metal.

Tempering of water-quenched welds

The heating at 650°C. of previously quenched welds is accompanied by further segregation of N-phase from the solid solution. Segregates oriented along the austenite crystallographic planes are observed at various stages in the heating. Further heating causes the segregation particles to coalesce and the oriented segregates to disappear.

The presence of oriented segregates was observed in quenched weld metal containing as little as 0·25% Nb after 1000 hr of heating. In this case, the niobium was introduced into the weld metal *via* the electrode coating; the formation of this type of segregate could therefore have been caused by local increase in the niobium concentration, since no such segregates were present in a weld containing 0·7% Nb even after 1000 hr of heating. It may be conjectured that these segregates are the N-phase, which separates from the solid solution at the comparatively low tempering temperature of 650°C.

A characteristic feature of the microstructure of welds containing over 2% Nb after tempering for 1000 hr is the large number of spot segregates, which are most probably N-phase. Austenite grain boundaries were not revealed even after holding for 1000 hr.

X-ray structural analysis of the precipitate of a weld containing 3·05% Nb after tempering revealed only N-phase and carbide of niobium. The formation of N-phase during prolonged tempering is accompanied by reduction in ductility and impact strength. In welds containing over 2% Nb the formation of N-phase after tempering so reduces the impact strength that the effect of the water-quenching is practically nullified.

Discussion

Together with the metallographic investigations, the X-ray structural analysis showed that where free niobium is present in welds various phases may form: the carbide of niobium, the niobide Fe_3Nb_2, N-phase, and the silicide $Fe_4Nb_5Si_3$. It is incontestable that the carbide–nitride phase Nb(C, N) forms in every case where the weld contains niobium. The phases Fe_3Nb_2 and $Fe_4Nb_5Si_3$, which are revealed where the niobium content is very high (over 2%), are formed in the welding process. Both phases are unstable, and are evidently dissolved by various heat-treatments, giving place to N-phase, which is stable even at high temperatures and the formation of which is linked with a high niobium concentration.

The fact that N-phase may form during normal heat-treatment—hardening or stabilizing—is extremely interesting. Possibly brief holding at high temperatures suffices for its formation. There is no doubt that the time interval during which the weld is in the high temperature region in normal two-pass manual arc welding is too short for N-phase to appear in the weld metal. Nor was N-phase found in any of the welds immediately after welding; it appeared in welds as a result of heat-treatment or prolonged tempering. Possibly N-phase is formed during the multi-pass welding of steels containing niobium, particularly where the niobium content exceeds its normal concentration.

The question of the temperature range in which N-phase forms, and that of the time required for its formation at various temperatures, are very important in the study of its formation in welds and heat-affected zones in austenitic steels containing niobium.

It may be conjectured that the process of the formation of N-phase is similar to that of the formation of sigma-phase, although it takes place within a wider temperature range at a higher rate, and is characteristic of steels with high niobium contents. The N-phase, like sigma-phase, is formed during prolonged tempering, mainly in the ferrite of the weld, but it also separates directly from the austenite. Segregates oriented along the austenite crystallographic planes are observed in welds with various niobium contents.

Where the Nb content is about 1·32%, the segregates appear only after 500 hr of tempering at 650°C.; where it is 3·05%, only 25 hr of tempering is required.

The segregates oriented along the crystallographic planes do not necessarily represent N-phase particles. Possibly N-phase is formed at the comparatively low temperature of 650°C. in several stages, the niobide Fe_3Nb_2 being first segregated along the austenite planes, and then converted into N-phase.

Increase in tempering time causes the coalescence and disappearance of the oriented segregates, unlike the case of the sigma-phase, where increase in the segregates is normally observed. Little is known of the crystallographic nature of the N-phase. Binder, as already noted, considered that it was an M_6C type carbide.[3]

Published data on the crystalline structure of the carbide Nb_3Cr_3C (M_6C type) do not correspond to values for the N-phase interplanar distances found by experiment. On the other hand, the χ-phase revealed in austenitic steels containing molybdenum corresponds very closely to N-phase.[14, 15] The distribution of the N-phase lines on the X-ray photograph corresponds to the distribution of lines for the χ-phase. Calculation of the N-phase lattice constant from experimental data, assuming its crystal lattice to be cubic as is that of the χ-phase, gave lattice constant values of $a_0=9{\cdot}40$ Å; thus, the ratio of the constants for N-phase and χ-phase ($a_0=8{\cdot}862$ Å) closely coincides with the ratio of the atomic radii of niobium and molybdenum. Further, the conditions under which χ-phase and N-phase form during many hours of heating at high temperatures are similar. The N-phase is formed in ferrite, as are the sigma-phase and the chi-phase. In the authors' opinion the N-phase should be considered to be an isomorphous χ-phase, in the same way as the Fe_3Nb_2 and Fe_3Mo_2 phases are isomorphous.

No extensive formation of N-phase would be expected in the heat-affected zone during the welding of type 18% Cr–11% Ni–B steel, since the niobium content of this steel is low. The N-phase is mostly formed in steels containing niobium in solid solution, or in austenitic alloys with high Nb contents.

The present research has not made it possible fully to establish the conditions under which other phases form (Fe_3Nb_2 and $Fe_4Nb_5Si_3$) or the conditions necessary for the conversion of one phase into another. With the small sizes of the X-ray structural analysis chambers the authors were not able to reveal these phases in the intermetallic precipitates which were taken from welds after 1000 hr of heating at 650°C. Evidently either the weld metal contains very small amounts of the niobide Fe_3Nb_2 and the silicide Fe_4Si_3, or they are mainly dissolved when the intermetallic precipitates are extracted. The formation of Fe_3Nb_2 or $Fe_4Nb_5Si_3$ in welds during welding is evidence that these phases, and even the eutectic of ferrite and Fe_3Nb_2, may form at high temperatures.

The formation of a eutectic from any of the phases enumerated above may be one of the causes of knife-line attack, a form of corrosion which develops in the narrow fusion zone. In this work an 18/8 type steel containing Ti was used. This did not permit the study of the formation of phases containing niobium in the fusion zone; the eutectic segregations observed in welds containing 1·32% Nb may be not only carbide type but may also be N-phase and Fe_3Nb_2.

Conclusions

(1) The alloying of 18/8 type welds with niobium reduces the ductility of the weld metal in proportion to the increase in the concentration of niobium.

(2) The following phases containing niobium form in welds in which over 2% Nb is present, in addition to the carbide of niobium: the niobide Fe_3Nb_2 and the silicide $Fe_4Nb_5Si_3$ (during welding), and N-phase (as a result of heat-treatment under various conditions).

(3) The presence of N-phase in the weld metal is accompanied by a reduction in ductility. The N-phase may form either in ferrite or in austenite. The crystal lattice of N-phase is isomorphous with that of the χ-phase which is found in 18/8 type steels containing molybdenum.

(4) Stabilizing-annealing or quenching cause the formation of N-phase in welds containing over 2% Nb. The formation of N-phase as a result of stabilizing annealing causes the impact strengths of weld metal to be roughly 20–30% lower than that of welds in the natural state.

(5) Prolonged tempering at 650°C. was accompanied by reduction in ductility. In welds containing less than 2% Nb this reduction in ductility is caused by the formation of sigma-phase. In welds containing over 2% Nb the reduced ductility is caused by the appearance of N-phase.

(6) Preliminary quenching has little effect on impact strengths after prolonged tempering of welds containing over 2% Nb. The impact strength proved to be extremely low, independent of whether or not quenching was carried out before prolonged tempering.

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Determination of Total Hydrogen Contents in Weld Metal

As a result of tests carried out in three laboratories, tentative recommendations have been made for the determination of the total hydrogen content in single beads of weld metal. A suitable form of standard test piece assembly is shown.

PARALLEL testing of one low-hydrogen and one rutile type weld metal specimen was arranged in three laboratories:

Murex Welding Processes Ltd. (Great Britain)
British Welding Research Association (Great Britain)
SINTEF (Norway).

The investigation was undertaken to test the compatibility of various methods of analysis with a view to establishing a recommended procedure for future work within the Sub-Commission.

The original analytical procedures were modified in various ways during the course of the work. In their present states, all three methods are considered equivalent with regard to analytical precision, but minor divergences are found owing to differences in the preparation of specimens. The recommended sampling method is intended for use as a conventional basis of reference. It is *not* implied that hydrogen contents obtained with this method of sampling actually represent the concentrations prevailing immediately after solidification of the weld metal.

Testing Procedures and Results

The electrodes used were commercial brands of 4 mm core wire dia.: *E* 455 *B* 11* of British and *E* 322 *R* 12 of Continental make, both supplied in sealed tins from the makers. Only one set of welding conditions was used, *i.e.*, manual welding at 170–180 amp and at a speed corresponding to about 1 cm of bead per 1·5 cm of consumed electrode.

The parent plate dimensions, cooling arrangement, number of layers, etc., differed from one laboratory to the other; experimental details have been published elsewhere.[1,2,3] Changes in procedure and apparatus made subsequent to the present investigation are described briefly at the end of the report.

Document IIW/IIS–8–58 (ex Doc. II–6–57) of Commission II—*Arc Welding* (Sub-Commission A—*Hydrogen in Weld Metal*) of the International Institute of Welding.

* IIW classification.

The final results are summarized in Table I. These figures are representative of the methods and equipment previously described.[1–3]

It should be noted that the figures of 'diffusible' and 'residual' hydrogen in Table I have, in the three laboratories, been measured in specimens of unequal sizes, and hence a comparison should be based on the sum of the two values. Furthermore, the 'diffusible' hydrogen collected from a single bead above glycerine at 45°C. may differ appreciably from the data given in Table I.

Discussion of Analytical Results

As might be expected, the hydrogen contents of deposited metal of multi-run deposits are considerably lower than those of single beads. In terms of fused metal (including the fused parts of the parent plate) there appears to be little difference between single-run and multi-run welds; however, this question will not be considered further in the present report.

The figures in Table I show that diffusible and residual hydrogen contents have been referred to different bases in the BWRA single-run tests. This, of course, tends to make the sum in these determinations too high in terms of fused metal and too low in terms of deposited metal. Taking into account the corrections that should be applied to these sums, the variance is not excessive. For the low-hydrogen electrode the deviation from the average of the Murex and SINTEF sums is $\pm 3\%$ for fused metal, and $\pm 6\%$ for deposited metal. Corresponding deviations for the rutile electrode data are $\pm 20\%$ and $\pm 9\%$, respectively. The rather large deviation found on a fused metal basis in this instance might be related to differences in penetration ratios. It is believed that differences in parent plate dimensions and in methods of specimen preparations will affect the final result more for a rutile than for a low-hydrogen weld metal.

The direct loading determinations tend to give lower results than measurements performed in two steps. With the exception of the SINTEF data for the rutile electrode, the total content found with direct loading is frequently lower than the 'diffusible' hydrogen measured in the same laboratory. One probable reason for this discrepancy is the loss of hydrogen to be expected in the earlier BWRA method, through

Table 1
Summary of comparative analyses of hydrogen in weld metal

Electrode	*Laboratory*	*Sample*	*Hydrogen content* ml/100 g *Single run* *Diffusible*	*Residual*	*Sum*	*Direct loading*	*Multi-run* *Diffusible*	*Residual*	*Sum*	*Direct loading*
Low-Hydrogen	Murex	Fused metal	1·63	1·65	3·3	—	0·94	0·84	1·8	2·6
		Deposited metal	3·54	3·6	7·1	—	1·24	1·1	2·3	—
	BWRA	Fused	—	0·06	<3·1[a]	—	—	—	—	—
		Deposited	3·07	—	>3·1[a]	2·81[d]	—	—	—	—
	SINTEF	Fused	3·0–2·6[b]	0·5–0·8[b]	3·5–3·4[b]	2·8[c]	—	—	—	—
		Deposited	4·9–5·1[b]	0·9–1·8[b]	5·8–6·9[b]	4·4[c]	—	—	—	—
	Murex	Fused	20·9	1·6	22·5	—	24·4	3·8	28·2	27·6
		Deposited	56·0	4·3	60·3	—	35·2	5·5	40·7	—
Rutile	BWRA	Fused	—	2·82	<41·9[a]	—	—	—	—	27·2[d]
		Deposited	39·03	—	>41·9[a]	33·07[d]	—	—	—	—
	SINTEF	Fused	28–30[b]	4·6–4·0[b]	33–34[b]	33	—	—	—	—
		Deposited	39–48[b]	6·5 6·3[b]	46–54[b]	53	—	—	—	—

(*a*) Diffusible and residual hydrogen not referred to same basis
(*b*) Range of hydrogen at locations 30 and 15 mm from end crater
(*c*) Measured about 45 mm from end crater
(*d*) Water vapour not included

reaction with an oxide skin in the sample. It must be admitted, however, that it is extremely difficult to remove the last traces of moisture from a chilled sample, and hence an unknown fraction of the water vapour recovered from such samples might actually originate from surface moisture. It is considered that the direct-loading procedure should not be recommended except for special purposes.

Another source of error in the direct-loading method might be connected with a reaction between dissolved carbon and the oxide skin. Irregular and excessively high gas contents (not shown in the Table) have at times been encountered, probably as a result of the formation of CO, and it appears absolutely necessary to determine the composition of gases extracted from directly-loaded samples. If, on the other hand, chemical analysis be omitted in routine determinations of residual gas, errors arising from CO contamination will probably be much smaller, because there will usually be no oxide skin present.

Differences in test-piece assembly and sample preparation may probably account for some of the variations shown in the tabulated results. Although the three methods are probably equivalent as regards reproducibility, it is considered to be more convenient to apply water quenching than cooling in carbon dioxide snow. Of the two water-quenched specimens used in this investigation, the smaller one[3] appears best suited for use with equipment available in other IIW countries.

It has been tacitly assumed that vacuum extraction of mild-steel weld metal at 650°C. will expel residual hydrogen to an extent that is sufficient for all practical purposes; any additional residual hydrogen to be recovered in vacuum fusion or in extraction at 1050°C. is assumed to be negligible in comparison with the scattering of results that must be accepted in determinations of this kind.

No general answer has been found to the question of how to report the total hydrogen content, and it appears advisable to present the analytical result in terms of both deposited and fused metal.

Conclusions

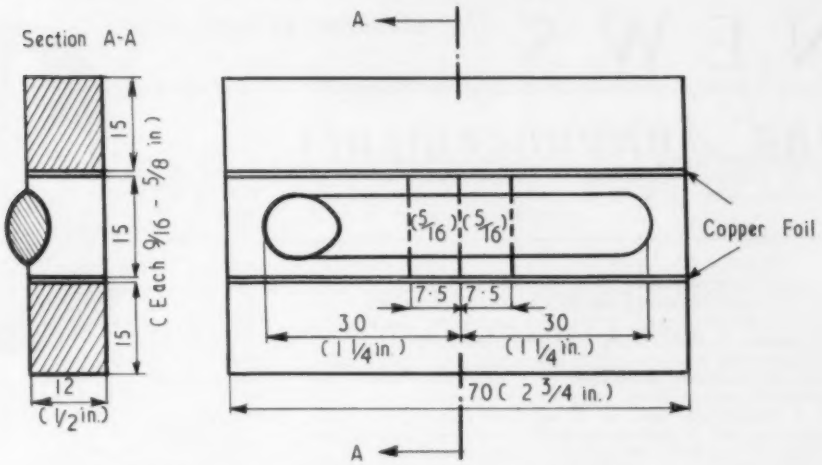
On the basis of the results and discussion presented here, a tentative recommendation for the determination of the total hydrogen content of single beads is as follows:

(1) A test piece assembly as shown in Fig. 1 is prepared from mild steel. The two central pieces used in the analysis are ground lightly and weighed to the nearest mg. For use with low-hydrogen electrodes these pieces should be degassed before grinding and weighing. The assembly is clamped by means of a toggle clamp or other quick release device and is placed on a mild-steel support.

(2) A single bead is deposited along the centre-line of the assembly, using the conditions recommended by the manufacturer, and a speed of welding corresponding to about 1 cm of bead per 1·5 cm of consumed electrode.

(3) The weldment is quenched in water immediately after completion of welding, and subsequently in alcohol saturated with solid carbon dioxide.

(4) The sample pieces are broken apart and wire brushed, successively washed briefly in alcohol and in ether, and dried in a blast of dry air or nitrogen. This operation should be completed within 30 sec.



1—*Recommended test piece assembly*

(Dimensions in mm, with nearest inch equivalents in parentheses)

(5) The samples are introduced into gas burettes of conventional design[1–3] and are left floating on mercury at room temperature until the evolution of gas has become negligible. It is advisable to transfer the gas to the main vacuum apparatus for the final measurement of cold extracted gas. To obtain the most precise result, the amount of hydrogen in these gases should be determined by means of a palladium osmosis tube.[1–3]

(6) The specimen is removed from the burette, weighed, and ground lightly so as to remove any visible oxide skin. It is then introduced into the hot extraction apparatus and baked at 650°C. until the evolution of gas has ceased.

(7) The hot extracted gases are analysed by a suitable method to discover their hydrogen content.[1–3]

(8) The combined volumes of 'diffusible' and 'residual' hydrogen divided by one hundredth of the difference of weight is the conventional content of total hydrogen in ml/100 g of deposited metal. The combined volumes of hydrogen divided by one hundredth of the weight of fused metal (as computed from the two cross-sections and the average length of the sample bead) is the conventional hydrogen content in ml/100 g of fused metal.

(9) The 'diffusible' hydrogen as measured in this procedure should not be compared directly with other conventional figures; *e.g.*, with those obtained using the method considered by Sub-Committee 3 (Filler Metals and Electrodes) of ISO/TC 44 (Welding).

APPENDIX

Changes in Procedure and Equipment after completion of Comparative Testing

Murex Method[1]

The present apparatus differs largely from that used previously in that the analytical section now includes not only:

(i) A Pirani gauge for continuous observation of pressure measurements

(ii) A side tube for freezing out the condensible gases CO_2 and H_2O

(iii) A palladium tube for direct determination of hydrogen by diffusion out of the system at 600°C.

but also:

(iv) A platinum filament for combustion of carbon monoxide at 550°C.

(v) A silver tube for admission of oxygen at 700°C. for the above combustion.

The method of analysis is the well-known low-pressure analysis technique developed by Ransley. It is therefore possible to analyse for carbon dioxide, carbon monoxide, water vapour, hydrogen, methane, and nitrogen.

BWRA Method[2]

(*a*) Hot extraction is carried out at 650°C. instead of 750°C.

(*b*) All welds are deposited on degassed mild steel channel

(*c*) A cylinder is turned from the fused zone of specimens after removal from **Y** tube collecting apparatus and before loading into hot extraction apparatus. (Formerly, the specimens were loaded without any kind of machining.) The hydrogen content obtained in this way thus cannot be converted to the basis of deposited metal because the whole of the 1 in. test piece is used to turn the cylinder for residual hydrogen determination. (Possibly, the relative amounts of deposited and fused metal may be estimated from sections through the run-on and run-off pieces.)

(*d*) Separate means have been provided for determination of water vapour.

REFERENCES

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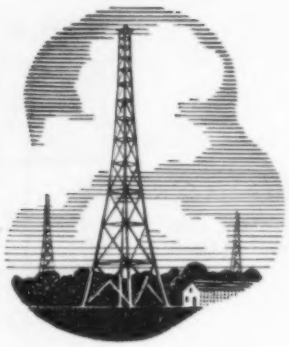
2. J. Walker: Report RFM/58—FM.3/110, British Welding Research Association, July 1955.
3. N. Christensen, K. Gjermundsen, and R. Rose: *Brit. Welding J.*, 1958, vol. 5, pp. 272–281.

NEWS
and Announcements



News of the Institute and Branches
B.W.R.A. · Other Societies ·
and Industry

INSTITUTE ACTIVITIES

MEETINGS

Second Annual Lecture

The Second Annual Lecture will be given at the Institute, 54 Princes Gate, London, S.W.7, on Thursday, **12th March 1959,** at 5.30 p.m. for 6.00 p.m.

The Lecturer will be Sir Leonard Owen, Managing Director of the United Kingdom Atomic Energy Authority, and his subject will be 'Welding in Relation to the Atomic Energy Programme'.

Admission will be by ticket only, obtainable from the Secretary of the Institute. In the first instance only one ticket will be issued to each member applying, but requests for additional tickets will be placed on a waiting list.

Spring Meeting, 1959

The Institute's Spring Meeting of 1959, to be held in London from **20th to 24th April** during the Engineering, Marine, Welding, and Nuclear Energy Exhibition at Olympia, will be the outstanding welding function of the year in Great Britain.

The notice, programme, and reply forms will be circulated to members in the third week of February, and it is expected that large numbers will take advantage of the opportunity to combine a visit to the Exhibition with attendance at some of the attractive Works Visits and important Technical Discussions forming part of the Spring Meeting.

An important social engagement will be the opening Reception at the Institute on the evening of Monday, 20th April. Those who will be in London on the following Wednesday night will also be able to obtain tickets for the Joint London Branches Dinner at the Connaught Rooms.

The last day of the Meeting, Friday, 24th April, is also that selected for the official visit of the Institute to the Exhibition, for which complimentary tickets will be issued with the Spring Meeting notice. The Council of the Institute will be the guests of the Exhibition Organizers at Luncheon at Olympia on that day.

SCHOOL OF WELDING TECHNOLOGY

A list of the Courses arranged for the early part of the year was published in the September 1958 issue of the Journal. This has now been extended to cover Courses to the end of September 1959. The programme is as follows:

D.15 **9th–13th February 1959:** Brazing Technology and Design
D.16 **23rd–25th February:** Welded Storage Tanks
D.18 **2nd–4th March:** Metal Spraying
D.6/1 **16th–20th March:** Welded Design and Construction in Corrosion- and Heat-Resisting Materials
D.8/1 **6th–10th April:** Welded Structures
D.9/2 **18th April:** Health and Safety in Welding
D.2/1 **11th–15th May:** Welded Design and Construction in Aluminium
D.7/1 **25th–27th May:** Ultrasonic Inspection
D.14 **8th–12th June:** Aircraft and Aero-Engine Welding
D.3/1 **22nd–26th June:** Welded Design and Construction in General and Mechanical Engineering
D.17 **21st–25th September:** Advanced course 'Welding Engineers

Brazing Technology and Design, 9th–13th February, 1959

Brazing offers to industry a most versatile method of fabrication by means of which it is possible to make many articles which would be difficult or uneconomic to produce in any other way. It is, however, a complex subject; nearly every metal may be used as filler material, and all methods of heating may be applied. Before the technical and economic possibilities of the process may be fully appreciated, it is necessary to have a proper understanding of the fundamentals of brazing, the selection of filler metals and heating method, and the correct approach to design. The School is therefore organizing a special Course with a view to giving up-to-date information about the wide range of brazing processes available, and guidance on their application.

There are still a number of places available on this Course.

SYLLABUS

	Lecturer	*Subject*
1	Dr. G. L. J. Bailey	Introductory Review
2	Mr. A. Cibula	Principles of Wetting Spreading and Capillary Flow
3	Mr. D. R. Milner	Principles of Oxide Removal
4	Mr. M. H. Sloboda	Brazing Alloys in Relation to Parent Metals
5	Mr. E. V. Beatson	Survey of Heating Methods
6	Mr. N. H. Jones	Principles of Design
7	Mr. G. F. Charge	Torch Brazing
8	Mr. R. Dickson	Salt Bath Brazing
9	Mr. J. B. Carroll	Furnace Brazing
10	Mr. T. J. Kirk	High-Frequency Brazing
11	Mr. J. Hinde	Brazing of Stainless and Heat-Resisting Materials
12	Mr. L. Bernhardt	Brazing of Copper
13	Mr. C. E. Densem	Brazing of the Less-Common Metals
14	Mr. E. Ryalls	Bronze Welding
15	Mr. F. Daintith	Inspection of Brazed Assemblies

NEWS OF MEMBERS

▶ The following awards are included in the New Year Honours List:
J. A. Milne (Kt.Bach); **Dr. Maurice Cook** (C.B.E.); **I. H. Hogg** (M.B.E.)

J. A. McWilliam, a member of Council and of the Executive Committee, has recently been appointed Chief of the Research Liaison Department of Firth-Vickers Stainless Steels Ltd., in succession to the late Mr. Edwin Smith.

J. H. N. Thompson, Deputy Chairman and Joint Managing Director of John Thompson Ltd., has been appointed a Director of Nuclear Chemical Plant Ltd.

▶ The Council regrets to record the death of **Lady Larke** on 10th January, and extends its deep sympathy to Sir William Larke and his family.

ELECTION OF MEMBERS

The following elections were made at the Council meeting of 27th November 1958:

Members

***A. Atkinson** (Berkhamsted); ***G. Gillam** (Lincoln); ***M. D. Jackson** (Sheffield); ***A. M. Kapadia** (Bombay); ***J. C. Kyte** (Weston-s-Mare); **J. McLean** (Motherwell); ***R. Pattison** (Colombo, Ceylon); ***R. Robinson** (Stockton-on-Tees); ‖**W. Stock** (Selby); ‖**A. G. Weller** (Nottingham).

Associate Members

†**W. N. Aspinall** (Rugby); **G. H. Batten** (Manchester); **R. J. Beishon** (London); ‡**F. F. Booth** (Oxon.); †**C. Boswell** (Bolton); **J. Bradshaw** (Manchester); **C. R. Breeds** (Ashford); **S. A. Budd** (Birmingham); **J. Bundred** (Carshalton); ‡**D. A. Burrell** (Kingston-on-Thames); **W. Calderbank** (Gainsborough); **L. V. Caron** (Montreal); ††**R. Chappells** (Rochester); ‖**W. A. Clapham** (Ellesmere Port); **J. C. Colling** (Bristol); **J. F. Corbett** (Transvaal, S. Africa); ‡**F. E. Darling** (Twickenham); †**A. A. L. Davies** (Teddington); **C. N. Dodsley** (Alfreton); **H. S. Doggett** (Streatham); **D. T. Downing** (Fulwell, Sunderland); **D. N. Evans** (Droitwich); **J. M. Fairgrieve** (Birmingham); **M. J. Feit** (Ashford); ††**A. Fenney** (St. Helens); ‡**E. A. Foreman** (Liverpool); †**B. G. Ghoshal** (Bengal, India); **G. P. Grainger** (Whitley Bay); **W. K. Hall** (Orpington); †**W. I. Halse** (Wimbledon); **C. C. Hanson** (Walsall); **K. Hemingway** (Wakefield); **M. L. Henry** (Kenya); **R. W. Holden** (Bolton); **W. Hughes** (Bexleyheath); **P. W. Hyde** (Sherbourne); **J. Irvine** (Esher); **H. J. John** (Cardiff); ‡**R. S. Lodey** (Cambourne); **H. Orrell** (Bolton); **W. Owen** (Newport); **R. W. Pannett** (Sidcup); **L. Plummer** (Bolton); **R. Purkayastha** (Calcutta); **E. W. S. Reed** (Aylesbury); †**K. Sharples** (Preston); †**D. F. Shaw** (Harrogate); †**J. M. Shaw** (Birmingham); †**F. E. Stock** (Woking); ‡**J. T. Stockdale** (Preston); **K. R. Sundaram** (Calcutta); †**J. F. Tobin** (Mold, N. Wales); **J. R. Watson** (Bristol); †**D. E. Wetherley** (Radlett); †**D. H. Woodhead** (Halifax); **H. Wormington** (Birmingham).

Companions

C. H. Bentley (Glasgow); **J. G. Bolton** (Dublin); **J. T. Courlander** (Ware); **J. G. Morton** (South Shields); **J. T. Oxborough** (Rotherham); **J. H. Pattenden** (Woking); ‡**G. Payne** (Greenford, Middx.); **E. J. Sansome** (Brit.W.Africa); **L. G. L. Watkinson** (Sudbury); **E. Wilson** (Sleaford); **J. Wood** (Oldham).

Graduates

R. J. Ibbotson (Sunderland); **W. J. Lyden** (Birmingham); **K. N. Mehrota** (Calcutta); **B. Mukherjee** (Calcutta); **D. R. Patterson** (Newcastle-upon-Tyne); **P. R. Richards** (Reigate); **C. J. A. Wood** (Birmingham).

Associates

A. D. Baxter (Northfleet); **L. F. Beckerley** (Deri, Glam.); **C. T. Butler** (Manchester); **R. F. Carrington** (Enfield); **W. Chamberlain** (Wolverhampton); **T. R. Cook** (Doncaster); **G. W. Cosgrove** (Peterborough); **R. M. Craig** (Paisley); **S. Cushing** (St. Pauls Cray); **T. R. Das Gupta** (Calcutta); **J. A. Dawson** (Sheffield); **W. S. Delamare** (London); **W. H. R. Derbyshire** (Upminster); **A. Gilbertson** (Accrington); **D. W. Gordon** (Glasgow); **C. A. H. Gough** (Bideford); **R. S. Gray** (Belvedere); **J. N. Hammerson** (London); **J. Harrison** (Birtley); **L. Hesketh** (Liverpool); **L. J. Iles** (Bristol); **P. Jayaraman** (Bombay); **F. Keating** (Oldham); **K. S. Krishnamurthy** (Bombay); **T. N. Lewis** (London); **J. R. Mainland** (Birmingham); **T. K. McEwan** (Paisley); **M. Milligan** (Darlington); **J. Minto** (Warminster); **N. J. Norvall** (London); **R. A. Ogunsolu** (Ibadan); **E. Partington** (Stockport); ††**J. B. Paterson** (Glasgow); **T. J. Pearce** (Helston) **D. B. Phillips** (Port Talbot); **R. A. Phillips** (Enfield); **R. K. Roy** (Bengal); **K. G. Sandry** (Swansea); **F. E. Shepherd** (Crawley); **H. S. Steel** (Bristol); **K. T. Stones** (Farnworth); **A. Thomas** (Leeds); **D. J. Waters** (Billericay); **B. Wordsworth** (Sheffield); **T. E. Worthington** (Cardiff).

Industrial Corporate Members

F. W. Berk & Co. Ltd. (Schori Division); **British Railways** (Eastern Region); **British Railways** (Scottish Region); **Brown Lenox & Co. (London) Ltd.; Corporated Engineers (India) Pvt. Ltd.; ††Govt. of New South Wales (Dept. of Railways); Joseph Adamson & Co. Ltd.; Metallizing Equipment Co. Ltd.; Tubewrights Limited.**

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* Transfer from Associate-Member
† Transfer from Graduate
‡ Transfer from Associate
‖ Transfer from Companion
†† Reinstatement

BRANCH NEWS—Reports of Meetings

East Midlands — Members' Papers

At a meeting held at Loughborough College on 18th November three short papers were presented by members of the Branch. The first, "*Bronze Welding by the Gas Process*", was given by Mr. C. E. Cobb, who gave a brief account of how 'Sifbronze' originated from the German alloy 'Tobin Bronze' and its further development in Britain. He considered that the term 'Bronze Welding' was rather misleading and that it would be better replaced by the description of 'Brazing'.

The process has advanced considerably in recent years and a number of bronze rods has been introduced from time to time. Yet experience has shown that it is difficult to improve upon existing standards of quality and strength.

Silver soldering was also described by Mr. Cobb, as a process used at a temperature between that needed for brazing and that for joining with hard lead solder.

Specimens made by bronze and silver solder brazing were exhibited.

"*Welding of Plastics*" was the subject of the second paper, introduced by Mr. A. Seaton. He confined his remarks to fundamentals, with particular emphasis on the welding of polythene. Plastics could be welded by the heated tool, by high-frequency, hot gas, and friction processes, but only the hot gas system was considered in the paper.

The special blowpipe used for welding produces a stream of heated gas at a temperature between 250° and 300°C., the melting point of polythene being about 110°C. Joint types are similar to those used in general steel fabrication – lap, fillet, butt, and corner joints. In butt joints, a square edge preparation is suitable for thicknesses up to and including $\frac{1}{4}$ in., but for greater thicknesses a vee preparation to give a 60–80° included angle and a root gap of $\frac{1}{32}$ in. is used. The filler rods are of the same grade as the material being welded. The welding technique is not difficult to master, but it is slower than in metal welding, and any attempt to increase the speed unduly produces poor results. The angle of the rod for butt joints is almost vertical and in line with the direction of welding. The blowpipe precedes the filler rod and the direction of welding is left to right. The rod is kept continuously pressed down on to the work.

Mr. Seaton ended his paper with a description of the application of plastics welding in the chemical industry; these included piping, tank liners, stirring rods, hoods, and fan casings. Some interesting specimens were exhibited.

The final paper, by Mr. L. Padmore, dealt with "*Some Aspects of Welding in the Chemical Industry*". The speaker stated that almost every welding method was used in fabricating equipment for the chemical industry. There was also a great variety in the materials needed, and a typical day's work might include the welding of aluminium, bronze, stainless steel, cast iron, mild steel, nickel, and various alloys for hardfacing.

Mr. Padmore then quoted two examples to illustrate the opposite extremes in size of job that he had experienced. The first was on a large construction in 1 in. thick copper plate, where two teams, each of two oxy-acetylene welders and one preheater, were employed on a 'one team on – one team off'

principle, and where a complicated jigging system had to be used.

The second example was the butt welding of 18 s.w.g. nickel sheet by the argon-arc process. For this the fit-up had to be virtually perfect, and the welding conditions had to be set to very fine limits. K.H.E.

East of Scotland — Metallurgy

For the fourth meeting of the session, held in November, there was a very good attendance of members to hear Mr. L. M. Gourd lecture on Introductory Metallurgy. In spite of the complexity of the subject the welcome given to the lecturer showed that there is a great interest in this most important aspect of welding technology.

Mr. Gourd introduced his subject with lantern slides and gave a detailed explanation of the various forms of grain structure in metals and their characteristics. This was followed by a general discussion, and the lecturer was congratulated on the easy and fluent way in which he had dealt with such a complicated subject. A.W.M.

Liverpool — Gas Cutting Progress

The 10th Annual Meeting between the Liverpool Metallurgical Society and the Liverpool Branch of the Institute of Welding was held on 12th November in the International Library Gallery of Picton Hall. A paper on "*The Gas Cutting of Metal Plates*" was presented by Dr. L. C. Bannister, and it interested a large gathering of metallurgists and welders from Merseyside.

Dr. Bannister referred to a lecture given by Thomas Fletcher to the Liverpool section of the Chemical Society in 1888, at which he demonstrated a gas-heated appliance for cutting metal. After a method of making acetylene from calcium carbide had been developed by the French chemist Le Chatelier a commercial burning torch for combined heating and cutting was used in Britain for the tapping of blast furnaces.

The lecturer then described the applications of the gas cutting process to various metals, including flame gouging, and cutting by the tungsten-arc process. He referred particularly to the great increase in the use of flame cutting in the shipyard and mentioned different methods of automatic and semi-automatic profiling.

Questions after the end of the paper ranged from triple flame cutting of plate edges for welding to the relative costs of cutting with different gas mixtures, and their variety demonstrated a keen interest in this all-important process in welded fabrication. H.R.C.

Sheffield — Corrosion- and Heat-Resisting Steels

Mr. J. A. McWilliam was the lecturer at the meeting held on 8th December, his subject being "*Argon-arc Welding of Corrosion- and Heat-Resisting Steels*". He illustrated his remarks with slides and showed also a number of exhibits, including a specimen of material 0·004 in. thick which had been welded by this process.

The meeting was well attended by about 80 members and visitors, and many interesting questions were ably answered by Mr. McWilliam. J.S.W.

South London — Structural Welding

Members and visitors heard an interesting lecture on "Some Aspects of Structural Welding" by Mr. E. M. Lewis at the Brixton School of Building, S.W.4, on the 26th November.

Mr. Lewis discussed several unit constructions which had been expressly designed to suit welded fabrication, and in each case he was able to show how, during the design of the component, its function had firstly been clearly understood, and that secondly the causes of failure of the earlier models, or perhaps failures of the existing ones, was known. Design, he pointed out, must not only take into account the cost but also safety. The examples he chose were buffers on a crane rail, the grandstand in a sports ground, and a special truck designed to carry loads at high temperatures and to receive loads which were dropped from a considerable height into it.

The lecture was most interesting and stimulated considerable discussion. E.G.B.T.

South Western — Welding Forum

The meeting on 4th December at Gloucester was held jointly with the Gloucester Association of Engineers. It took the form of a "*Welding Forum*" with Mr. W. K. B. Marshall, Mr. R. J. Burt, Mr. W. A. Woollcott, and Dr. C. L. M. Cottrell as Panel members, under the Chairmanship of Mr. F. J. Wilkinson.

The questions put to the Panel were answered very competently, and this type of meeting aroused considerable interest in the area. It is likely to become an annual event in the Branch programme. G.F.T.

Annual Dinners

No less than four Branches held their Annual Dinner during the first week of December.

On 2nd December the East Midlands Branch held their Dinner at the Victoria Station Hotel, Nottingham. Mr K. H. Ewen, Chairman of the Branch, presided, and the Toast of the Institute and the East Midlands Branch was given by the Branch President, Mr H. L. Palmer; Mr John Strong, the President of the Institute, responded. The Secretary of the Institute, Mr G. Parsloe, proposed the Toast of the Guests, and the response was made by Mr Robert Jenkins, J.P., Past President of the Institute.

The speakers at the Leeds and District Branch 16th Annual Dinner at the Mansion Hotel, Roundhay, Leeds, on 4th December, were Councillor Albert King, Deputy Lord Mayor of Leeds, who proposed the Toast of the Institute, Mr. John Strong, who responded, and the Branch President, Mr. A. E. Rose, and Mr. W. G. F. Sutcliffe, Managing Director, Richard Sutcliffe Ltd., who respectively proposed and responded to the Toast of the Guests. The Branch Chairman, Mr. R. Gardener, presided.

On the same night the Birmingham Branch held their Annual Dinner at the Grand Hotel, Birmingham. The Chairman was Mr. A. L. Hale, Branch Chairman, who also proposed the Toast of the Guests. The Toast of the Institute of Welding and the Birmingham Branch was proposed by Sir Arthur Smout, J.P., and the response was made by Mr. R. E. G. Weddell, Past President of the Institute. Mr. R. Hampshire responded for the Guests.

The North Eastern (Tyneside) Branch held its Annual Dinner at the Old Assembly Rooms, Newcastle upon Tyne, on Saturday, 6th December, with Mr. W. R. Mellanby, Branch Chairman, in the Chair. The Toast of the Institute was proposed by the Chairman and the response was made by Mr. Robert Jenkins. Mr. C. Stephenson proposed the Toast of the Guests to which the response was made by Mr. G. H. R. Towers, a Vice-President of the North East Coast Institution of Engineers and Shipbuilders. Mr. L. Cooley, Chairman of the Tees-side Branch, proposed the health of the Chairman, to which Mr. Mellanby briefly responded.

OTHER SOCIETIES

American Welding Society—Annual Meeting

The 40th Annual and Spring Technical Meeting of the American Welding Society, and the Committee on Electric Welding Conference of the American Institute of Electrical Engineers is being held in the

Guests at the East Midlands Branch Dinner
Left to right: E. A. Everritt, A. Robert Jenkins, J. Strong, K. H. Ewen, H. L. Palmer, G. Parsloe, N. L. G. Lingwood.

Hotel Sherman, Chicago, Illinois, on 6th–10th April 1959.

The 7th A.W.S. Welding and Allied Industry Exposition is also being held in conjunction with the meeting, in the International Amphitheatre, Chicago, on 7th–9th April.

In line with previous meetings there will be technical sessions for the discussion of papers on almost every phase of welding, demonstrations of new equipment and techniques, and visits to works.

New England Conference

The 3rd Annual Conference sponsored by the No. 1 District of the American Welding Society was held at Providence, Rhode Island on 23rd October. The Conference was attended by G. O. Hoglund, the Society's President, and the National Secretary, F. L. Plummer, and many members and visitors.

The three-part programme included a session on education, with three papers on Structural Welding, Welding of Tanks, Pressure Vessels, and Piping, and on Cost Cutting through Welding.

At another session modern techniques for welding stainless steel were demonstrated, as well as methods for welding piping and cast iron.

The final session was devoted to the discussion of papers on Inspection, Welding for Nuclear Service, and Brazing for High-Temperature Service.

After a banquet there was a talk by Roger Clark on Qualification and Training Welders.

NEWS FROM INDUSTRY

Thermit welding

Thermit Welding (Gt. Britain) Ltd., of Rainham, Essex, has been formed by Murex Ltd. and Elektro-Thermit G.m.b.H. of Essen, Germany, to pool the technical knowledge and experience of the two firms in the field of welding.

There has been a considerable amount of continuous welding of rails on the Continent in recent years, and the experience gained will now be available to operators in Britain and other countries. The new Company will also offer an improved service for rail and crane-rail welding, building up of worn rails, and the repair and construction welding of heavy machine parts.

Murex Ltd., who have already ceased to market independently Thermit products, will however supply compounds and spare parts to their specifications for a limited period for the benefit of former customers.

The Manager-Designate of the new Company is Mr. A. J. Hills, formerly Manager of the Thermit Welding Department of Murex Ltd.

English Electric welding equipment

The year 1958 has been one of progress for the Welding Division of the English Electric Co. Ltd. It has introduced a new range of weatherproof multi-operator transformers with built-in power-factor correction capacitors and fuse switches, for use in shipyards and the heavy engineering industries. This equipment, the first of its type, has been well received and is in great demand. Where it is used on site special units have been provided employing a tertiary winding to supply auxiliary circuits for hand tools and lighting.

A new portable double-operator welding set is meeting the demand for occasional heavy welding, and is also available with built-in power-factor correction capacitors.

With increasing interest in the use of rectified a.c. supply the new 25–350 amp. d.c. welding rectifier has proved to be popular in both home and overseas markets It takes a balanced 3-phase load at a high power factor and has an infinitely variable current control.

The range of electrodes has been expanded particularly in the iron powder types. These provide remarkable increases in the speed of welding and can be used satisfactorily by relatively unskilled operators. Four types are now available: 'Speedees' (general-purpose, rutile); 'Hermees' (low-hydrogen); 'Pyristees IP' (creep-resisting); and 'Pressurees IPL' (for pressure vessels and deep grooves).

The now well-established 'Groovees' electrode is being widely used for gouging out faults in welds, preparing backs of vees for sealing runs, and piercing holes in a variety of metals.

In its Liverpool laboratories the Company has been carrying out extensive creep testing of weld deposits under many different service conditions, and this experience is available for the formulation of electrodes to meet specific requirements.

Mobile inspection units

For some years mobile radiography units have been operated by Mitchell Engineering Ltd., for the inspection of Class I high-pressure piping for steam and central power stations. These units are now being increasingly used for welded work on nuclear contracts. The work involves the inspection of argonarc welded pipe connections in stainless steel, Monel, and Inconel, made in position at site.

Each unit is a large caravan trailer containing a film processing dark room, isotope handling devices, and a small workshop fitted with grinding and drilling machines, collapsible drawing board, and a 7-ton hydraulic bending machine for making bend tests.

Tube End Preparation

The use of simple portable rigs, which can be easily moved to any convenient place in the works, has enabled John Brown Land Boilers Ltd. to reduce cost and increase output in the preparation of tubes for butt welding at their Glasgow works.

Each rig consists of a Consolidated Pneumatic type 350–R–100 heavy-duty power vane drill mounted in line with two ball races and a vice on a common baseplate. A specially designed cutter head, carrying two Cleveland twist drill cutters of 10% cobalt steel, can deal with tubes from 1½ to 6 in. outside dia, and in thicknesses from 10 gauge to ½ in.

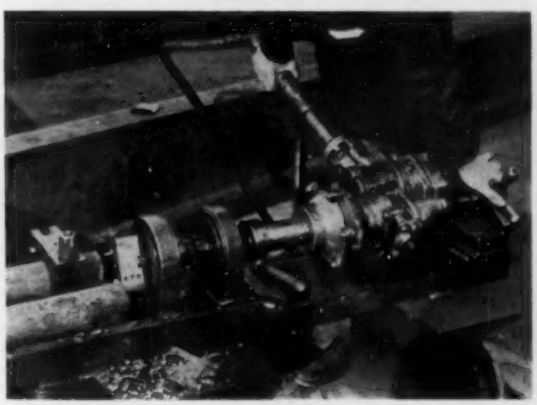
Although the rig enables cuts to be made within limits of 0·005 in., only semi-skilled labour is needed for its operation.

Electrical heat-treatment

Electrical resistance heating elements are becoming increasingly used for the preheating, stress-relieving, and general heat-treatment of welds because of their adaptability to flat or curved surfaces, their high heat capacity, and ease of control.

The application, performance, and method of operation of Electrothermal

Electrical resistance heating coils being used for preheating. Heavy thermal insulation protects the welder.



Preparation of boiler tube ends for butt welding, using portable rig

Armoured Heaters are concisely described in a new leaflet No. 321 issued by Electrothermal Engineering Ltd. This gives all the information needed (including prices) for deciding the type of installation suitable for a particular application.

Electrode Tip Dressing

A tip dressing tool for resistance spot welding electrodes has been produced by Portable Welders Ltd. The tool has two floating cutter blades, which automatically centre themselves on the two electrode tips, which can remain in the welding machine.

It is claimed that normal electrodes can be accurately hand dressed to shape in under a minute. Two models are available; the Industrial with a capacity up to $\frac{3}{4}$ in. dia., and the Junior for tips up to $\frac{1}{2}$ in. dia.

Welding Symbols

A new standard, AWS A2.0–58, on welding symbols has been issued by the American Welding Society. The increase in the use of welding and in the number of welding processes has meant that the number of symbols needed has doubled.

In the new standard symbols have been introduced for spot and seam welds made by arc welding, for welds in aircraft, guided missiles, and automotive work, and for structural steel fabrication, concrete reinforcement and the like.

A new system of designating welding and cutting processes by letters is included in tabular form, and the chart of welding symbols has been redesigned to incorporate recent developments.

The standard, and copies of the chart for wall or desk use, are available from the offices of the Society, 33 West 39th Street, New York 18, New York, U.S.A.

STUB ENDS

▶ An agreement has been reached between The General Electric Co. Ltd. and Vacuum Industrial Applications Ltd., which will combine the resources of the two companies for the development of vacuum heating techniques and the design and manufacture of complete vacuum furnace installations, including those for brazing.

▶ Dr. Thomas H. Osgood, of Michigan State University, has been appointed Science Officer in the United States embassy in London. Dr. Osgood, a physicist, received a Ph.D. degree from the University of Cambridge in 1927.

▶ A brief survey of non-destructive testing methods, and an introduction to ultrasonic testing are two of the articles in the December 1958 issue of 'SS News'.

▶ The share capital of Hancock & Co. (Engineers) Ltd. and also that of Weldcraft Ltd. have been acquired by the board of G. D. Peters & Co. Ltd. Mr. Hancock, who established his firm just forty years ago, will continue as Technical Director, and Mr. E. Seymour-Semper, who founded Weldcraft in 1944, will remain as Managing Director. The Director and General Manager of G. D. Peters & Co. Ltd., Mr. H. J. Stretton, will be the Chairman of the new board. The names of both companies will be retained.

▶ MAPEL (Metal & Pipeline Endurance Ltd.) have been awarded a contract valued at more than £50,000 for the inspection of the 26–30 in. pipeline from Gach Saran to Kharg Island in the Persian Gulf, which is being constructed for the National Iranian Oil Co.

▶ Preparation of materials and suitable techniques for lead welding are described in the Autumn 1958 number of 'Torch'.

▶ The multi-head 'Oxyplane' plate edge preparation machine, which allows the four edges of a plate to be cut to a variety of forms at one time, is described in a brochure recently issued by British Oxygen Gases Ltd.

▶ An illustrated catalogue (and price list) recently issued by D. S. Baddeley Engineering Co. Ltd., contains comprehensive details of a range of gas welding and cutting equipment, arc welding transformers and accessories, resistance spot and projection welding machines, and welding rods and fluxes.

▶ The Kent Refinery on the Isle of Grain, now dealing with nearly $7\frac{1}{2}$ million tons of crude oil a year, is described in a handsome booklet just published by The British Petroleum Co. Ltd.

DIARY

2nd Feb.—West Wales—*Recent Thoughts on Structural Welding* by A. V. Hooker (Swansea Technical College).

Manchester—*Weld Inspection* by D. T. Carter (College of Science & Technology).

3rd Feb.—East Wales—*Modernization of British Railways* by M. G. R. Smith and R. A. Smeddle (S.W. Inst. Eng., Cardiff).

East Midlands—*Weldability of* ND *Steels* by L. Reeve (Welbeck Hotel, Nottingham, 7.15 p.m.).

5th Feb.—Eastern Counties—*Metal Spraying Techniques* by Mr. Shepherd (City College, Ipswich Road, Norwich, 7.30 p.m.).

North Eastern (Tyneside)—*Planning for Welding in Shipbuilding* by P. H. H. Rhodes.

6th Feb.—South Western—*Gas Cutting and Welding* by L. B. Smith (Radiant House, Bristol 1, 7.15 p.m.).

9th Feb.—Sheffield—*General Considerations in the Welding of Materials for Industrial and Chemical Plant* by M. Birkhead (Grand Hotel, Sheffield, 7.15 p.m.).

10th Feb.—Liverpool—Demonstration of Non-Ferrous Gas Welding (College of Technology, Byrom Street).

12th Feb.—South London—*Welding and the Draughtsman* by H. B. Merriman (54 Princes Gate, S.W.7, 7.30 p.m.—Joint Meeting with **North London** Branch).

Eastern Counties—*Non-Destructive Testing Methods* by Mr. Shepherd (Colchester, 7.30 p.m.).

13th Feb.—Birmingham—*Welding in Nuclear Plant* by S. H. Griffiths.

Leeds—Dinner Dance (Mansion Hotel, Roundhay, Leeds).

17th Feb.—North London (Slough Section)—*Underwater Welding and Cutting* (Community Centre, Slough, 7.30 p.m.).

Leeds—*Welding in the Rocket Industry* by F. J. Wilkinson (Gt. Northern Hotel, Leeds, 7.30 p.m.).

18th Feb.—Wolverhampton—*Iron Powder Electrodes* by A. B. Fieldhouse (Holly Bush Hotel, Wolverhampton, 7.30 p.m.).

West of Scotland—*The Computer-Controlled Flame Profiling Machine* by C. Steventon.

North Eastern (Tees-side)—To be arranged.

25th Feb.—North London—*Brazing and Soldering of Small Articles* by L. D. Connell (54 Princes Gate, S.W.7, 7.30 p.m.).

25th Feb.—East of Scotland—*Metal Spraying* (North British Hotel, Edinburgh)

2nd March—West Wales—*Non-Destructive Testing* by H. Webborn (Swansea Technical College).

3rd March—East Wales—*Welding in the Rocket Industry* by F. J. Wilkinson (S.W. Inst. Eng., Cardiff).

Eastern Counties—*Metal Spraying* (Ipswich, 7.30 p.m.).

5th March—North Eastern (Tyneside)—*Shipyard Reorganization for Welded Construction* by R. Straton.

6th March—South Western—*The Self-Adjusting Arc Welding of Aluminium and Its Alloys* by E. M. Wilson (Radiant House, Bristol 1, 7.15 p.m.).

Birmingham—*Industrial Air Liquification and Rectification* by F. Din.

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All other matters relating to advertisements for this section of the News should be addressed to the Executive Editor, 6 Ridge End, Hook Hill Lane, Woking, Surrey (Tel. Woking 2981).

Copy should be sent by 6th of each month for publication in the following month.

Current WELDING LITERATURE

Book Reviews
Additions to the Institute Library
Trade Publications

CONTENTS OF PERIODICALS RECEIVED

This section is intended to give a survey of the current welding literature received by the Institute of Welding Library. The contents lists are not exhaustive; only the main articles in welding journals are listed, and reprints from other journals and short notes are generally excluded. In addition, welding articles from other periodicals are listed. Abstracts of welding literature are given in the Bibliographical Bulletin of Welding and Allied Processes, *published by the International Institute of Welding, and details of this may be obtained from the Secretary of the Institute of Welding.*

Welding Journals

Industry and Welding (U.S.A.), 1958, vol. 31, October

Will aluminium replace steel in highway bridges? (60–61, 90)
How stainless piping was welded for an atomic sub. (62–63, 98–99)
Techniques for fabricating magnesium (64–65, 96)
Resistance welding: a fast way to join dissimilar metals (66–67, 86)
Six joints to use when welding titanium (68–70)
Tungsten arc welding, F. Kuncz (71–72, 118–119)
Braze aluminium to stainless (73, 99)
Stored energy percussive welding joins steel studs to plated trim (74–75)
Cold working relieves weld stress, eliminates metal removal, H. L. Meredith (76–77, 86)
Flame cutting small parts cuts operating time 511%, W. Gallo (78, 92)
Mechanized valve surfacing with the oxy-acetylene flame, J. F. Barnes (81–82)

Welding Engineer (U.S.A.), 1958, vol. 43, October

A guide to nuclear power, Pt. 1—Principles, W. D. Briggs (36–38)
A.B.C.'s of weld estimating, C. F. Brown (39–41)
Geodesic steel dome is put to its first U.S. industrial use (42–43)
Resistance welding procedures and design, L. F. Spencer (44–47)
Filler metals for joining, O. T. Barnett (48–50)
Ultrasonics: a 'sound' welding technique (52–53)
Unusual design guarantees clear view of sports, E. A. Moenning (56–57)
A key to non-destructive test symbols (70–72)
Weld quality without guesswork, N. Nyberg (74–76)

Zvaranie (Czechoslovakia), 1958, vol. 7, September–October

The economical effect of welding technology in mechanical engineering, J. Cabelka (257–261)
Automatic electro-slag welding of steels 11 474 and 13 123 up to thickness of 120 mm, A. Kovar (261–270)
Automates for welding into molten slag developed in the Welding Research Institute in Bratislava, M. Mosny (270–274)
Some examples of economical application of welding into molten slag, J. Zeke (274–278)
Automatic submerged arc welding, M. Zaruba (278–282)
Some possibilities of the utilisation of automatic submerged arc welding, J. Stembera (282–284)
New experiences in carbon dioxide shielded arc welding, Z. Duben (284–286)
CO_2 shielded metal arc welding, A. Hrabovec (286–288)
Some problems about CO_2 shielded metal arc welding, L. Lanyi (288–289)
Argon shielded arc welding in the construction of steam generators, J. Novotny (289–291)
Welding of allied steels in the Lenin Works in Pizen, V. Pilous (291–297)
Some remarks to the question of embrittlement of austenitic steels type 18/8 and 16/13 Cr/Ni, A. Zapletalek (298–300)
Longer lifetime for ploughshares, J. Skriniar (300–303)
Gas shielded arc cutting with tungsten electrode, M. Matejec (303–307)
Friction welding, J. Mandaus and J. Vaboril (307–314)
Site welding, R. Kapoun (314–320)

Canadian Welder, 1958, vol. 49, September

Computor controlled profiling machines (14–16)
Brussels Fair 'Atomium' is all-welded (18)
Structural welding: inter-relation between materials, details of design, construction procedures, and service conditions, Pt. 1, Dr. LaMotte Grover (20–22)

Journal of the Japan Welding Society, 1958, vol. 27, September

On the influence of oxygen upon the consumptions of cathode and anode in arc, M. Ozawa, I. Morita and K. Omura (18–23)
On the porosity of weld steel with low hydrogen type electrode (Report 2), H. Sekiguchi and I. Masumoto (24–28)

Study on welding of low alloy high strength steels (Report 1), Y. Ito and Y. Araki (29–35)

Cooling time and brittleness of various steels submitted to welding (Report 3), H. Sekiguchi and M. Inagaki (36–40)

Hot cracking of austenitic chromium-nickel stainless weld metal with FISCO tester (Report 1), J. Maekawa, K. Matsuo, T. Suzuki, and S. Yamajo (41–47)

Comparison of corrosion resistance of welded commercial pure titanium and practical corrosion resisting alloys, Y. Ando, T. Fujimura, and M. Tobita (48–52)

On wettability and contact angle of soft solder, Y. Ishii and A. Kamio (53–58)

Schweissen und Schneiden (Germany), 1958, vol. 10, October

Automatic welding process with magnetically controlled electric arc, for the welding of pipe lines, H. Günther (385–394)

Flame gouging in production and repair work, H. H. Grix (395–400)

Economic aspects of the new German Standard DIN 4100 'Welded building structures, calculation and structural design', W. Wrycza (401–404)

Adaptation of gas cylinder fittings to suit automatic filling processes, F. Ninnelt (405–410)

The welding of screwed rings on cast iron pipe lines with screwed fittings, M. Jürgens (411–412)

Soudage et Techniques connexes (France), 1958, vol. 12, September–October

A welding process applicable to the fabrication of steel pressure vessels for atomic energy purposes: argon arc welding with a consumable electrode, J. Tigeot (325–335)

Welding developments in Austria, F. Rapatz (339–344)

Welding at the service of Lacq gas field workings, P. Berthier, R. Enous, J. Souviron, R. Terminet, and G. Reynes (357–364)

A comparison of methods of assessing spot weld ductility in sheet steels, E. J. French and A. A. Wells (369–392)

Other Journals

New power source broadens use for CO_2 welding, H. J. Bichsel, E. J. Steinert, and E. R. Gamberg (*The Iron Age* (U.S.A.), 1958, vol. 182, October 23)

The tinning of wire for soldering; advantage of hot-dipped coating (*Tin and its uses, Quarterly Journal of the Tin Research Institute*, 1958, No. 44, Autumn)

Health hazards in metal spraying, W. E. Ballard (*Metal Industry*, 1958, vol. 93, November 7, pp. 389–392)

Hardening of rail ends, G. K. Colligan (*New Zealand Engineering*, 1958, vol. 13, September 15, pp. 343–349)

How to surface shafts with stainless steel, R. G. Alison (*Machine Production*, 1958, vol. 17, October, pp. 19–20)

Chromium plating on sprayed metal, R. de Buyer (*Metal Industry*, 1958, vol. 93, October 31, pp. 373–374)

Sprayed high melting point metals, J. Cauchetier (*Metal Industry*, 1958, vol. 93, October 31, pp. 374–375)

Sprayed and diffused metal coatings, R. E. Mansford (*Metal Industry*, 1958, vol. 93, November 14, pp. 413–416)

Check testing welds and welders, L. A. Woodworth (*Engineer and Foundryman* (South Africa), 1958, vol. 24, September, pp. 57–62)

Magnetic flaw detection, A. G. Gardner (*Metallurgia*, 1958, vol. 58, November, pp. 257–259)

High strength titanium alloy has good weldability, F. A. Crossley (*The Iron Age* (U.S.A.), 1958, vol. 182, October 30, pp. 84–86)

Production line steps up girder welding, J. V. Banks (*The Iron Age* (U.S.A.), 1958, vol. 182, October 30, pp. 90–92)

Welding and brazing in atomic reactor construction (*Mechanical World*, 1958, vol. 138, November, p. 516)

Pressure vessels in nuclear plants, H. Allen (*Mechanical World*, 1958, vol. 138, November, pp. 517–519)

How to machine stainless steel—Pt. 5 (*Machine Production* (Canada), 1958, vol. 17, October, pp. 31, 60, 62)

Modern methods of fabricating structural steelwork, M. F. Palmer and R. J. Fowler (*Structural Engineer*, 1958, vol. 36, November, pp. 363–376)

Pressure vessels and the problem of brittle fracture, H. Harris (*Metallurgia*, 1958, vol. 58, November, pp. 213–217)

On the mechanism of propagation of brittle fracture in mild steel, M. Yoshiki and others; **The welding carried out in ship assemblies under vibrations produced by repeated impact**, T. Yoshida and others; **Studies on the effect of residual welding stress on brittle fracture of steel—Pt. 2**, H. Kihara and others; **Distribution of residual stresses in butt-welded joints**, K. Masubuchi (*Journal of the Society of Naval Architects of Japan*, 1958, vol. 102, February)

Joining and fastening of materials—Materials selector, 1958–59 issue (*Materials and Methods* (U.S.A.), 1958, vol. 48, mid-October, pp. 371–393)

ADDITIONS TO THE LIBRARY

BOOKS AND PAMPHLETS

L. J. ANTHONY, *Select list of standard British scientific and technical books*. 5th ed., 1957, Aslib, London.

INSTITUTION OF MECHANICAL ENGINEERS AND INSTITUTE OF WELDING. *Symposium on aluminium pressure vessels*, October 28th, 1958, Critical review of existing codes of practice for the construction of pressure vessels in aluminium and its alloys, R. T. Rose and J. G. Young, pp. 3—12; Properties of British aluminium-base materials suitable for the construction of pressure vessels, E. Elliott, pp. 13–28; Fabrication of aluminium pressure vessels, J. F. Lancaster, pp. 29–39; Experience with aluminium pressure vessels, C. A. Terry and G. B. Roberts, pp. 40–48. Institution of Mechanical Engineers, London.

Ryland's *Directory of coal, iron, steel, tinplate, metal engineering, foundry, hardware, and allied trades*. 33rd ed. 1958, Industrial Newspapers Ltd., London (Price 84*s*.).

Transactions of the Institute of Welding and Welding Research Supplement, Index to years 1939 to 1953. 1958, Institute of Welding, London (Price 25*s*.).

The BEAMA Handbook, 1958–1959. 1958, British Electrical and Allied Manufacturers' Association (Incorporated), London (Free on request).

API STD. 1104—*Standard for field welding of pipe lines*. 5th ed. May 1958, American Petroleum Institute, New York (Price 50 cents).

ROGER W. BOLZ, *Metals engineering processes*. (ASME Handbook.) 1958 (Price 104*s*. 6*d*.); Samuel L. Hoyt, *Metals properties*. (ASME Handbook). 1958 (Price 85*s*. 6*d*.). McGraw Hill, London.

BS. 1633:1958, *Steel for land boilers, receivers and other pressure vessels*. 1958. This revised standard deals with the plates, sections, bars, rivet bars, and rivets used in the construction of land boilers, receivers and other pressure vessels. An appendix gives the ratio of the 0·2% proof stress at elevated temperatures to the minimum specified tensile strength at room temperature. British Standards Institution, London (Price 6*s*.).

IS:813–1956. *Indian standard scheme of symbols for welding*. 1957 (Price Rs.3.00); IS:814–1957. *Indian standard specification for covered electrodes for metal arc welding of mild steel*. 1957 (Price Rs.3.00); IS:815–1956. *Indian standard classification and coding of covered electrodes for metal arc welding of mild steel and low alloy high-tensile steels*. 1957 (Price Rs.2.00); IS:816–1956. *Indian standard code of practice for use of metal arc welding for general construction in mild steel*. 1957 (Price Rs.2.00); IS:817–1957. *Indian standard code of practice for training and testing of metal arc welders*. 1958 (Price Rs.4.00); IS:818–1957. *Indian standard code of practice for safety health requirements in electric and gas welding and cutting operations*. 1958 (Price Rs.4.50); IS:819–1957. *Indian standard code of practice for resistance spot welding for light assemblies in mild steel*. 1957 (Price Rs.2.00); IS:1181–1957. *Indian standard qualifying tests for metal arc welders (engaged in welding structures other than pipes)*. 1958 (Price Rs.2.50). Indian Standards Institution, Manak Bhavan, 9 Mathura Road, New Delhi 1.

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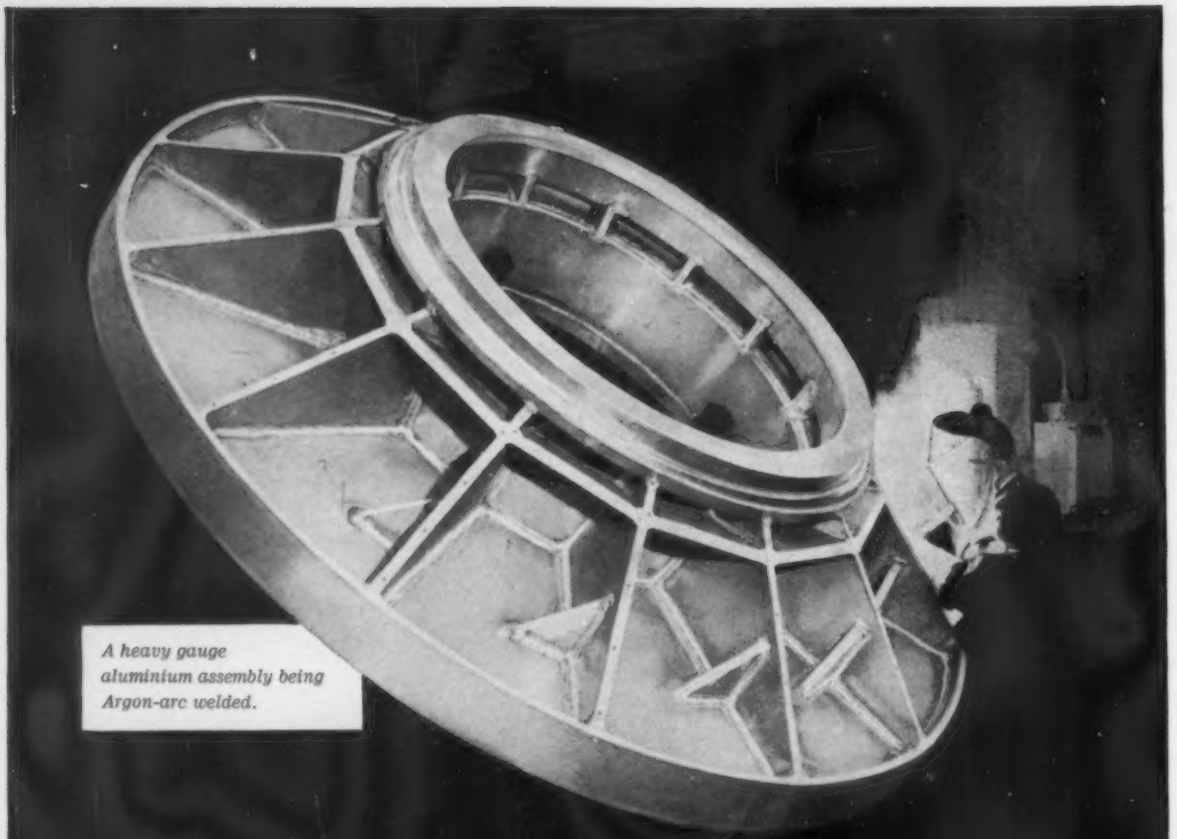

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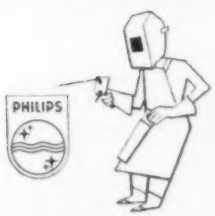

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